DECEMBER, 195

STACKS

JOURNAL OF THE IRON AND STEEL INSTITUTE

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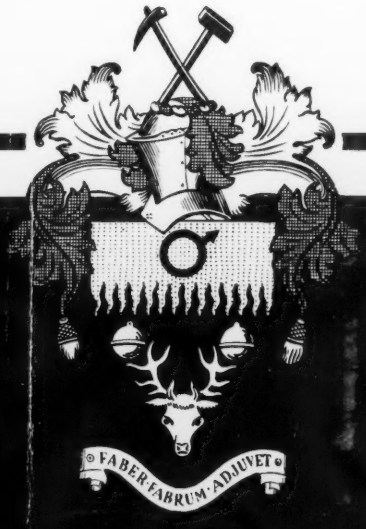

VOL. 187 PART 4

Principal Contents

Tenth Hatfield Memorial Lecture
by Paul G. Bastien 281

Low-carbon Bainitic Steels
by K. J. Irvine and F. B. Pickering 292

Sinter-bed Ignition *by H. Bates* 310

Tempering Low-alloy Creep-resistant Steels
by E. Smith and J. Nutting 314

The Tuyere Combustion Zone
by J. Taylor, G. Lonie, and R. Hay 330

Determination of Niobium and Tantalum
by the B.I.S.R.A. Methods of Analysis Committee 341

News 344 Abstracts 349

Translations 348

A full list of contents is given on the Title Page (see page 61, Advertisement Section)

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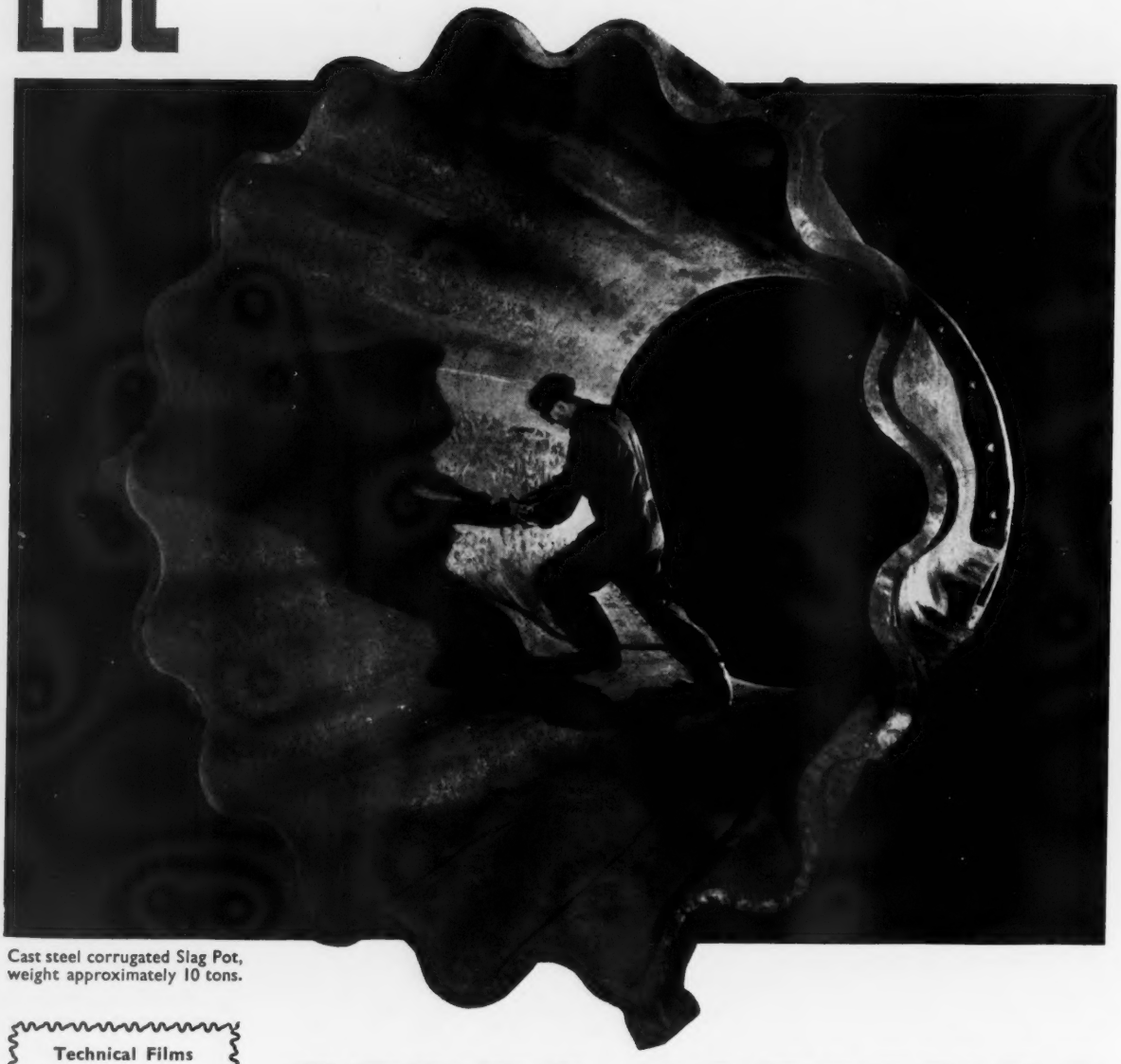

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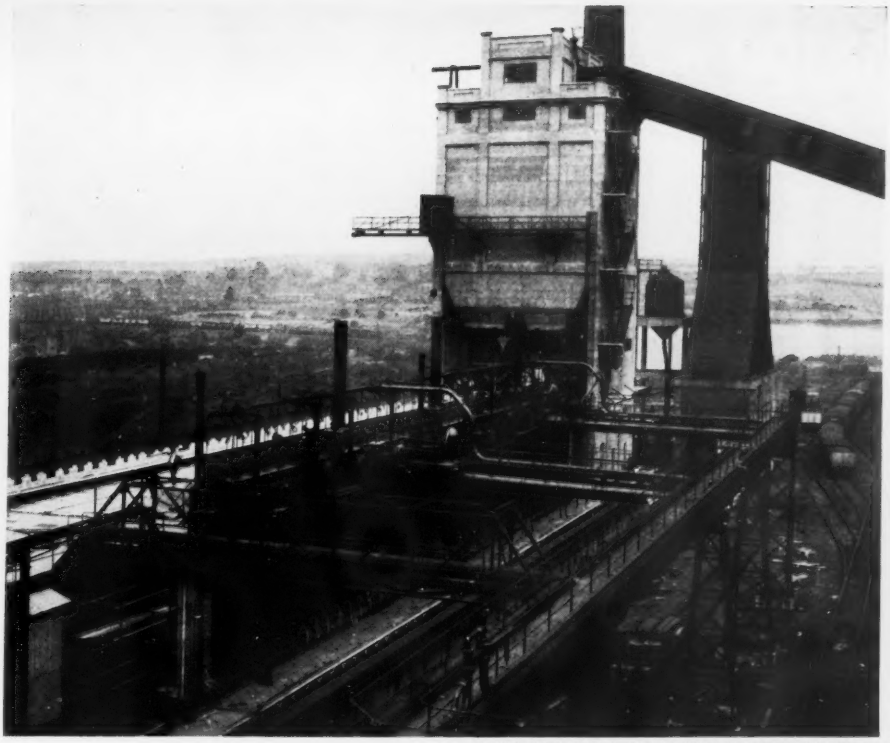

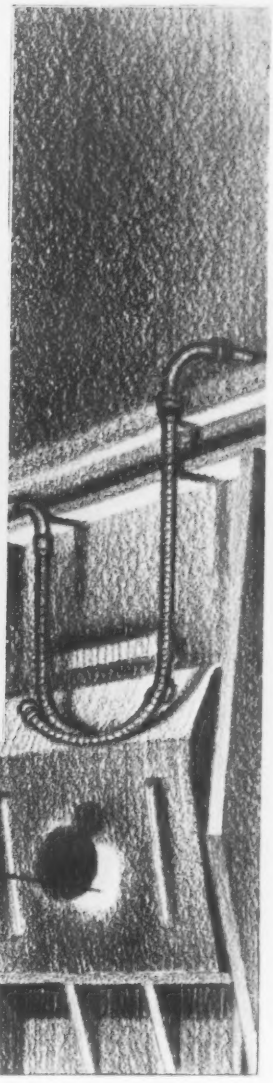

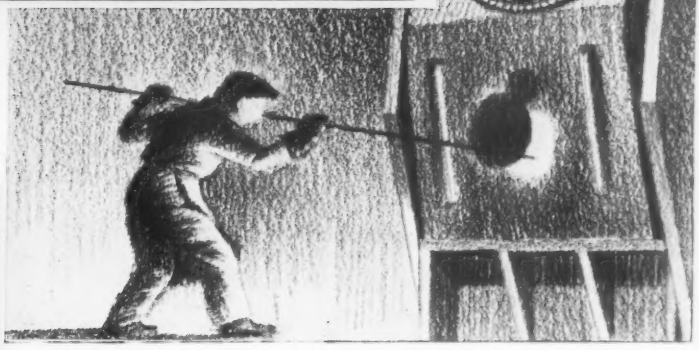

More ovens for more steel

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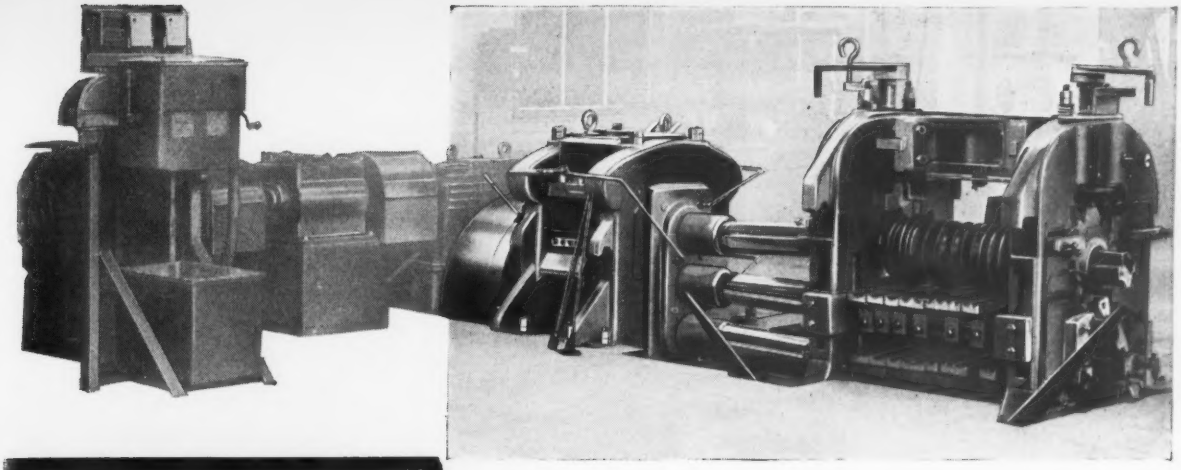

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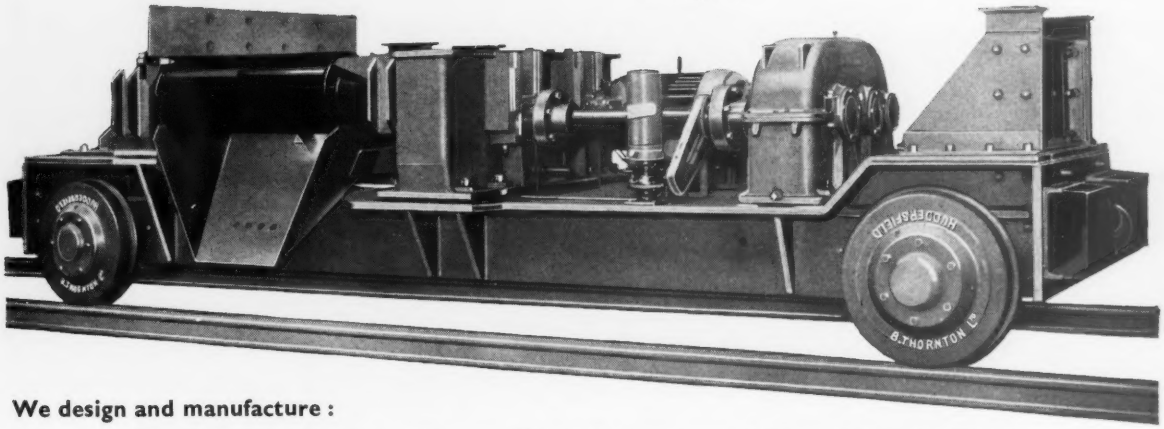

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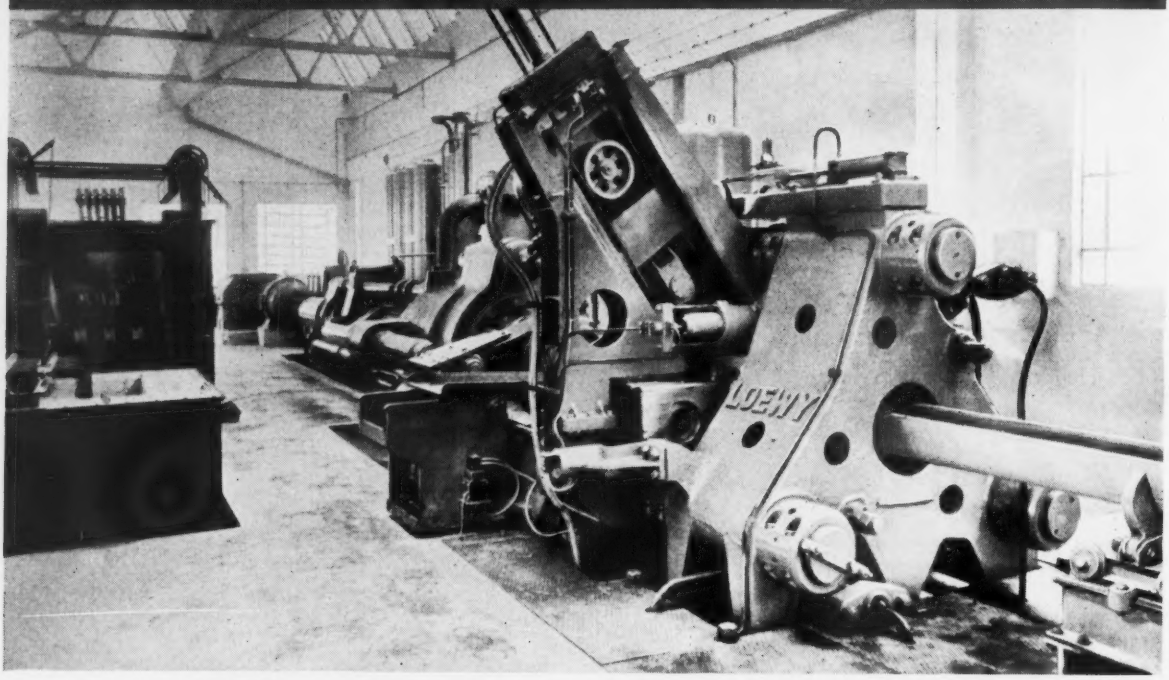

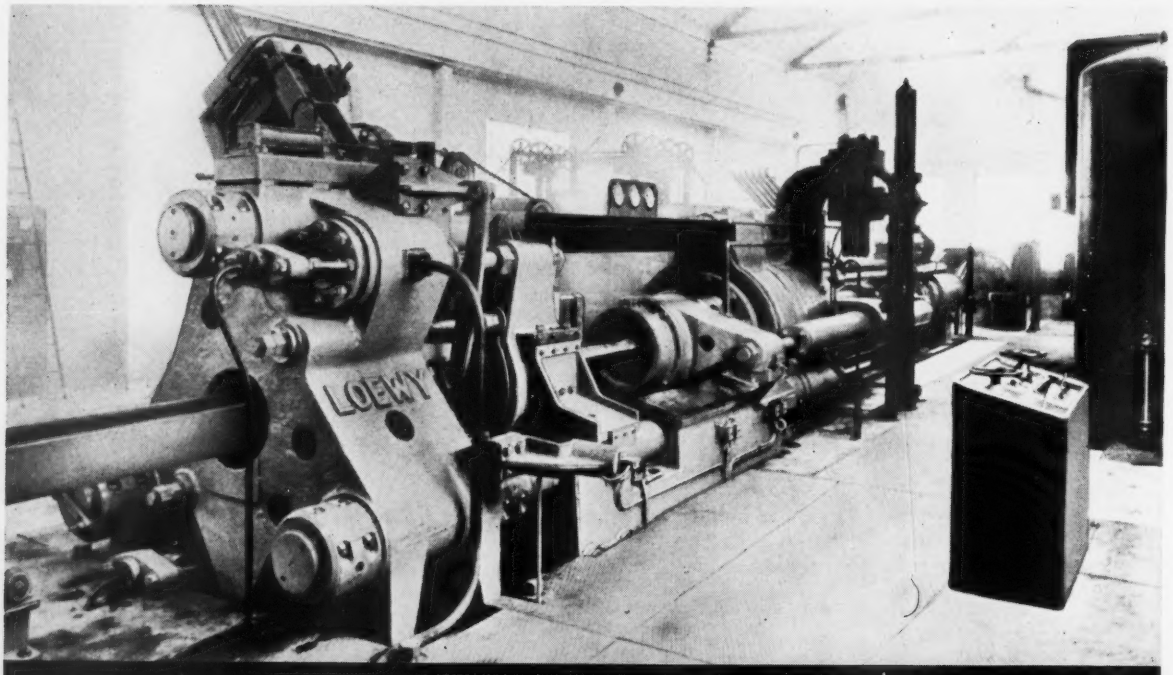

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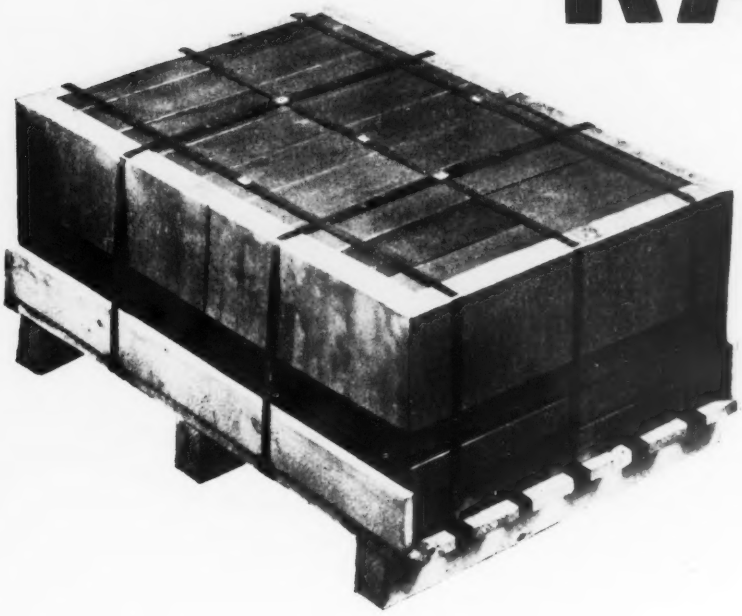

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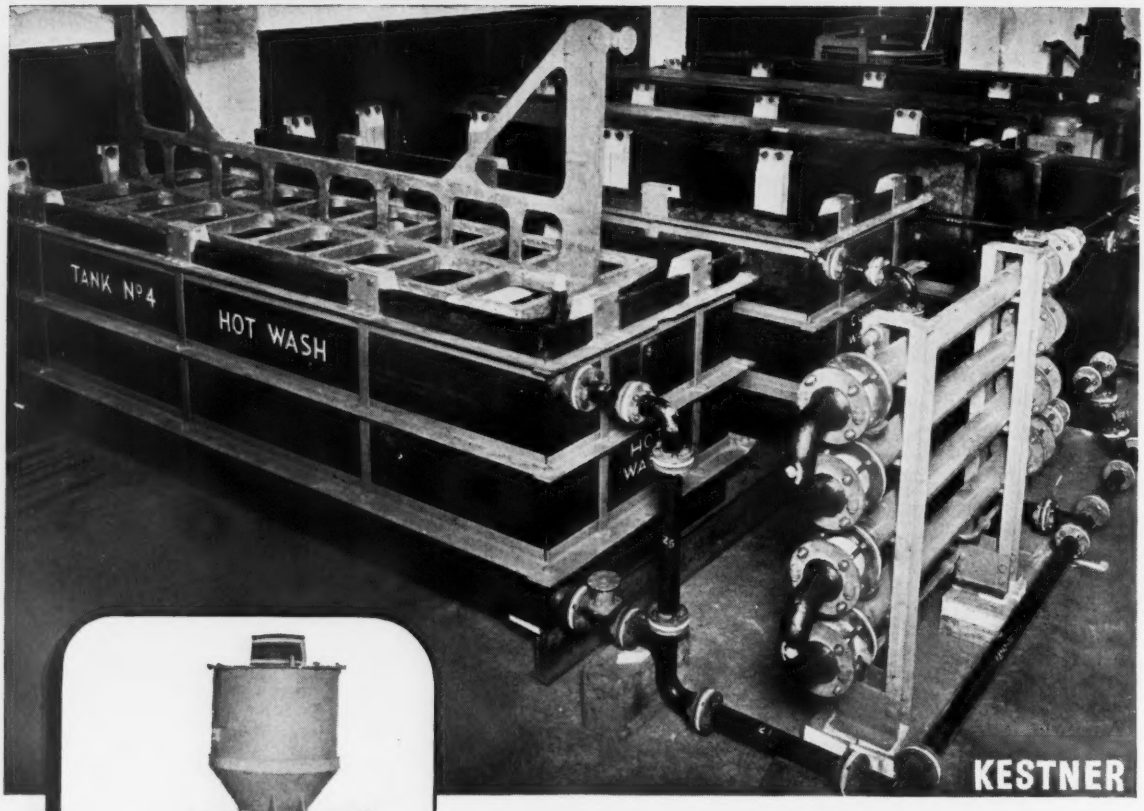

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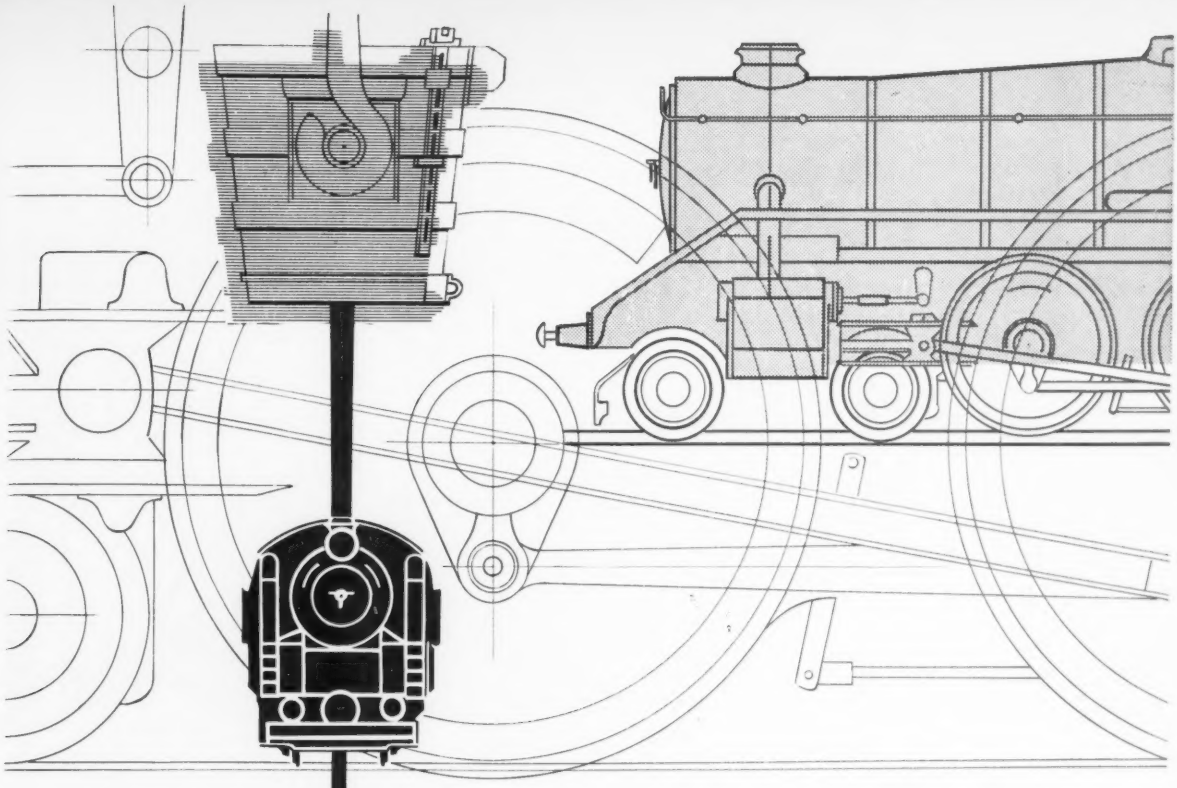

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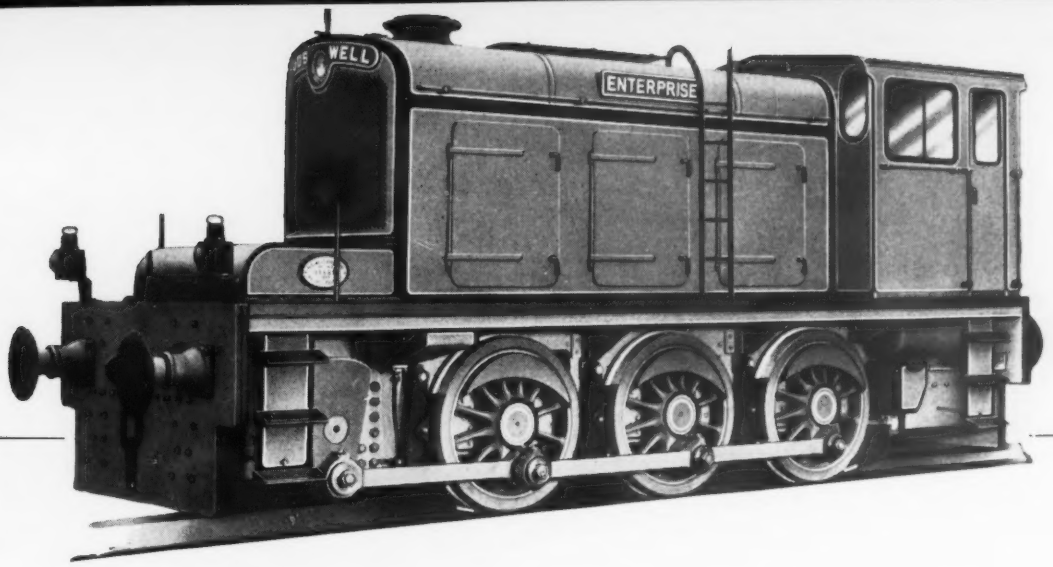

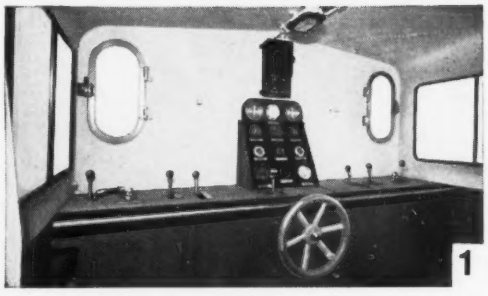
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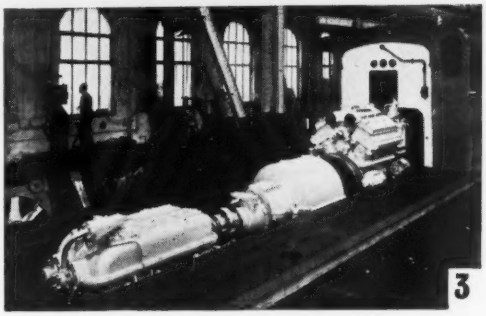
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1. *The cab layout. The central forward-and-reverse handle, with which is combined the high-and-low alternative ranges, is centrally located; the other controls are duplicated.*

2. *'Enterprise' at the head of a coal train*

3. *Installation in the locomotive 'Enterprise' showing left to right, the final reduction gearbox, the Dual Fluidrive transmission, and the Paxman V-6 pressure-charged constant h.p. engine.*

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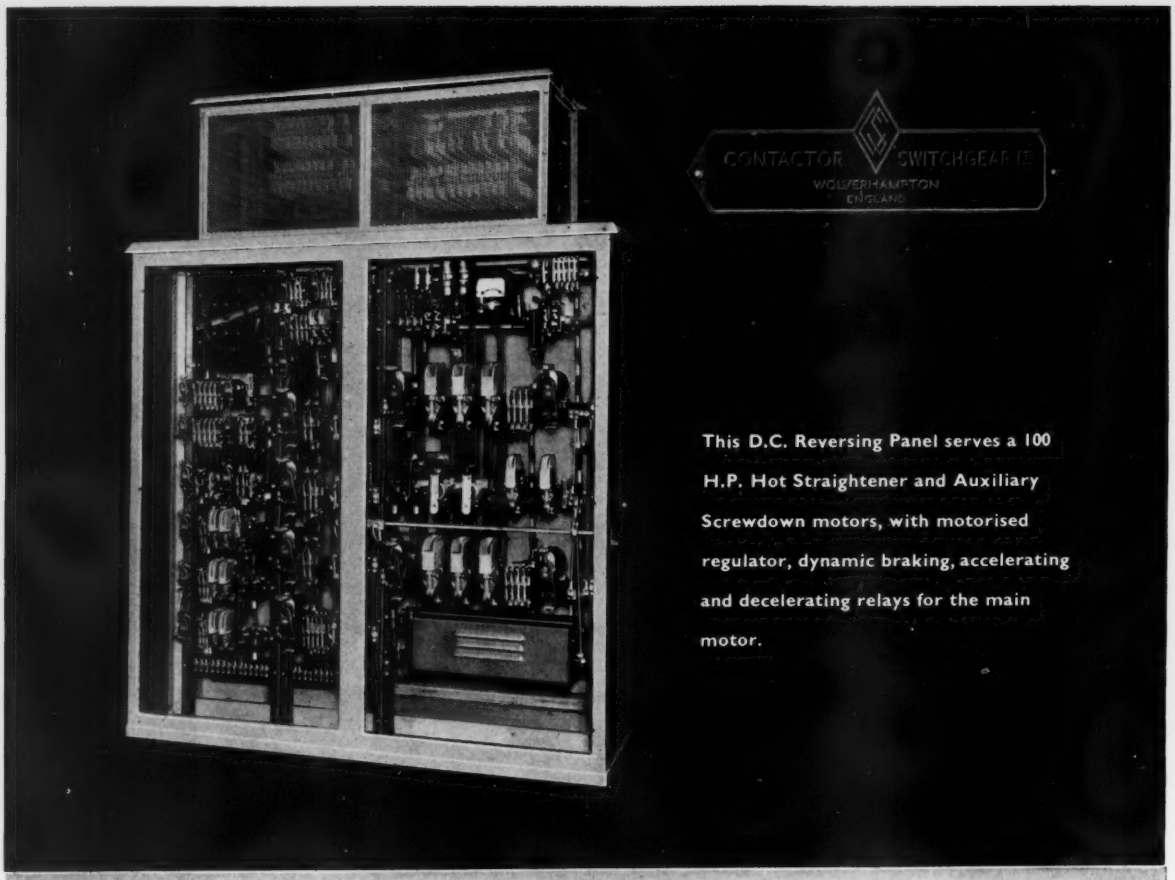

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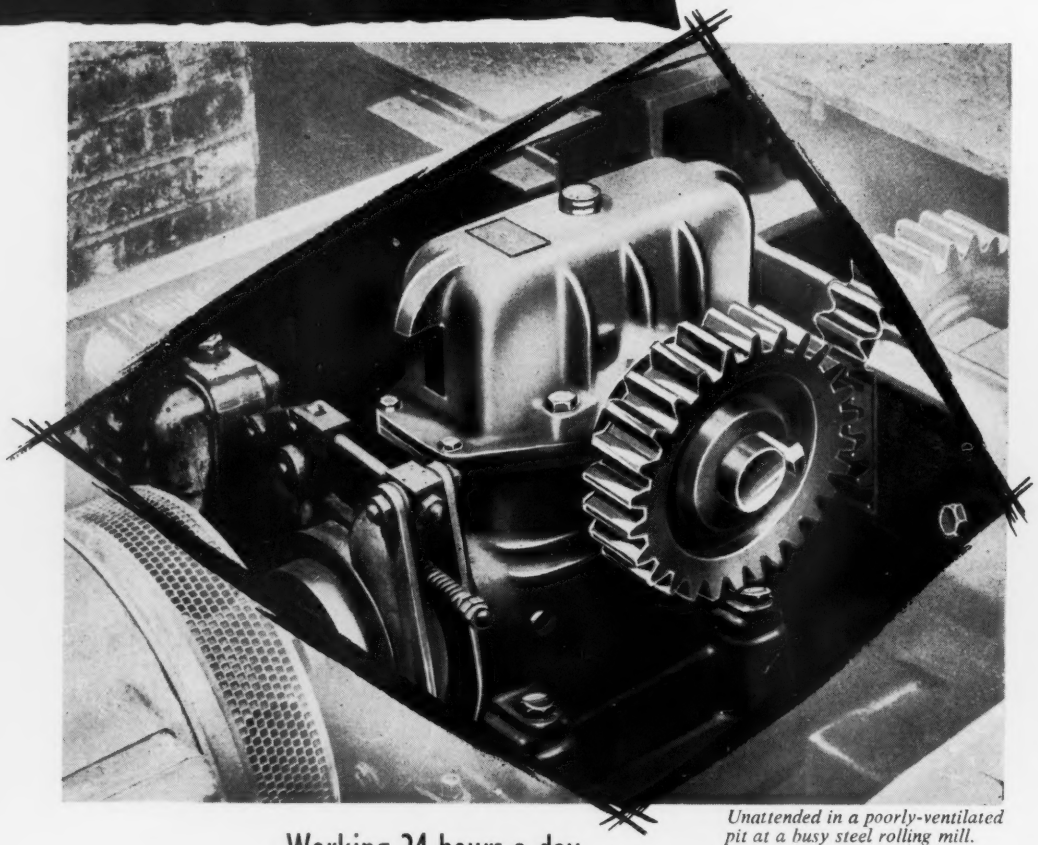

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to move 600 tons of hot steel a week
and averaging 130,000 stops and starts a month,
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LARGEST OF ITS TYPE IN OPERATION IN EUROPE

lined throughout with

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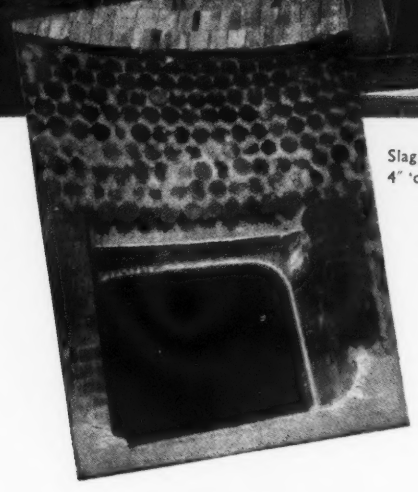

Slagging door showing 4″ 'overhang' of tubes

Furnace in 25° tilt position—maximum tilt 40°

Oughtibridge Magnesite Tubes were selected for lining the new furnace at Fabrique de Fer de Charleroi, Belgium as a result of their remarkable record of long and efficient service elsewhere.

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Pictures by courtesy of Fabrique de Fer de Charleroi and The Pittsburgh-Lectromelt Furnace Corporation.

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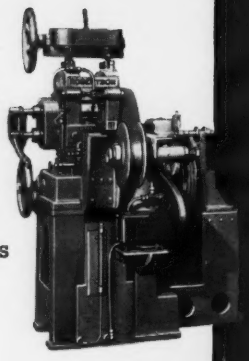
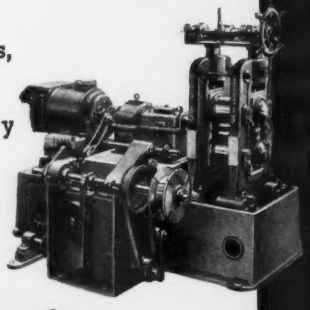
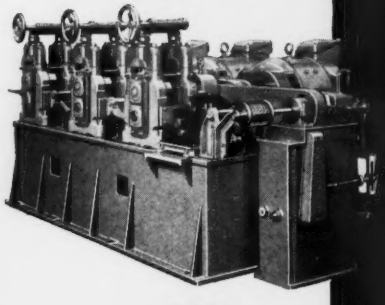


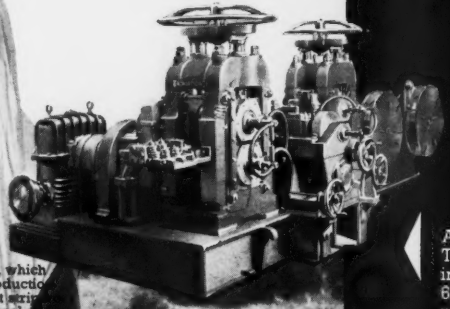
A small Two-High Mill with rolls 4″ diameter, 3″ face.



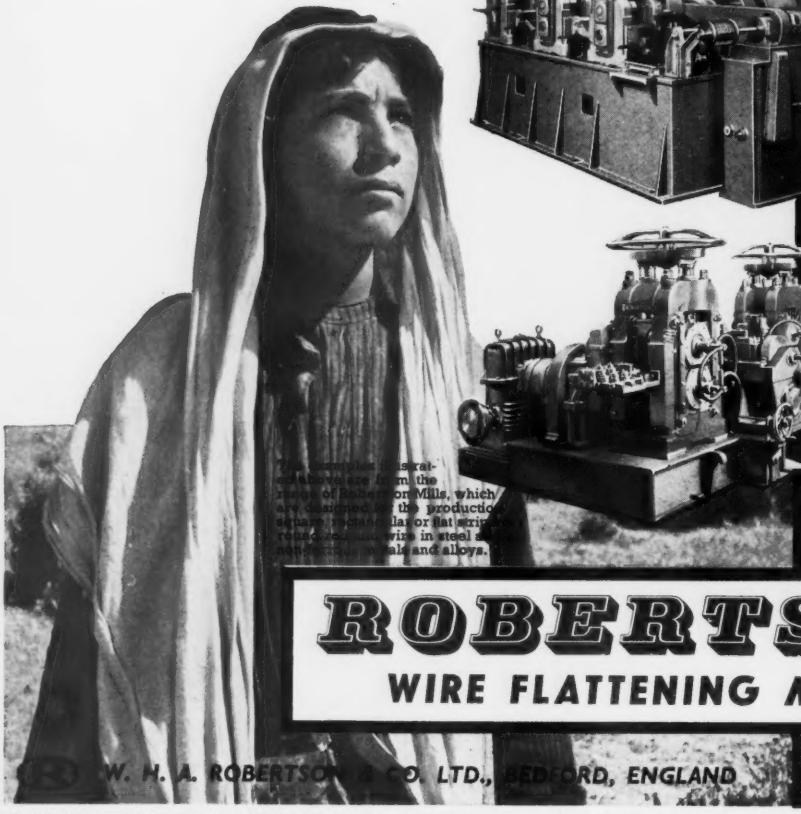
A Four-High Single Stand Mill with work rolls 2¼″ diameter, 4″ face, and support rolls 6¼″ diameter, 4″ face.



A Three-Stand Tandem, each stand individually motor driven and having rolls 6⅛″ diameter, 4″ face.



A No. 2 Tormanco Tandem, with roughing rolls 10″ diameter, 6″ face, and finishing rolls 8″ diameter, 6″ face



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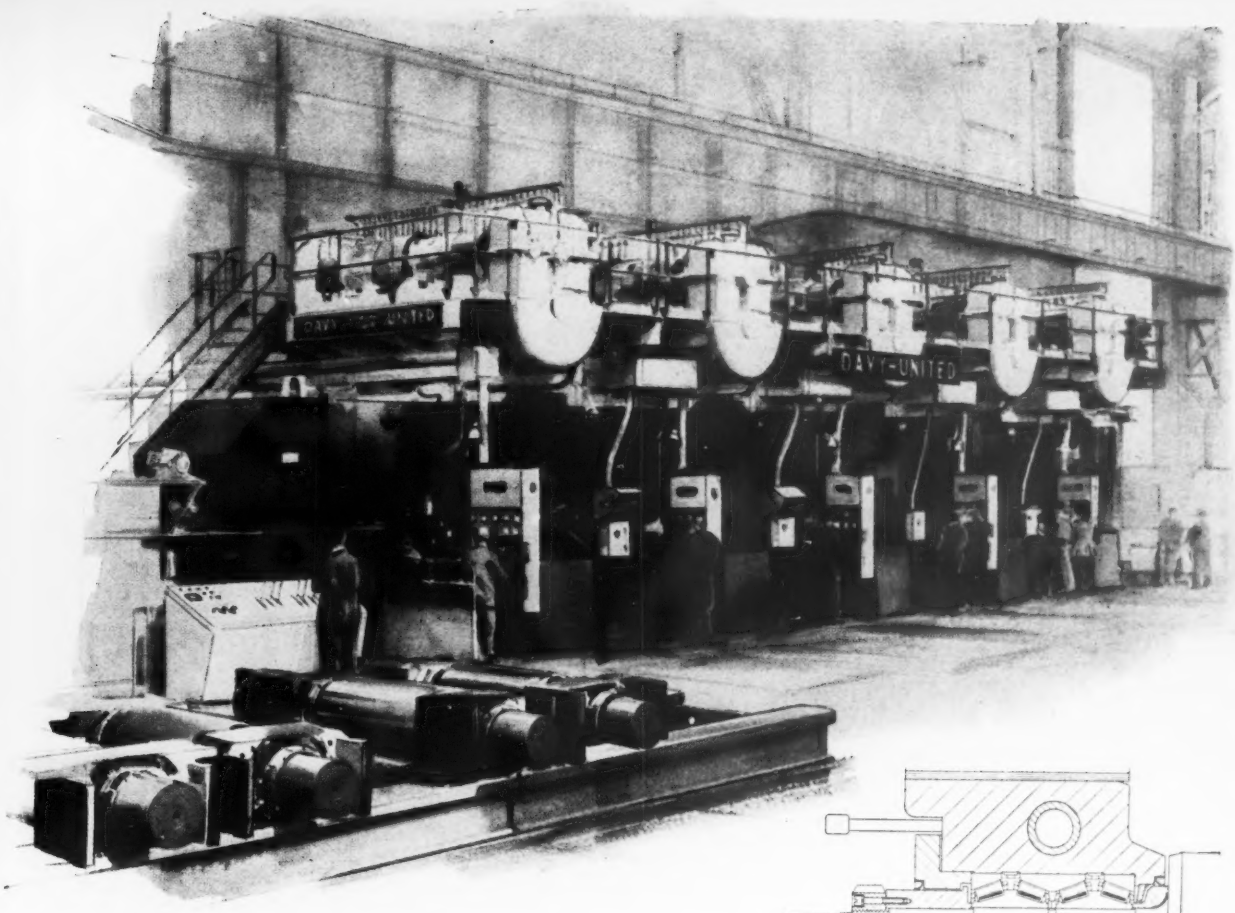
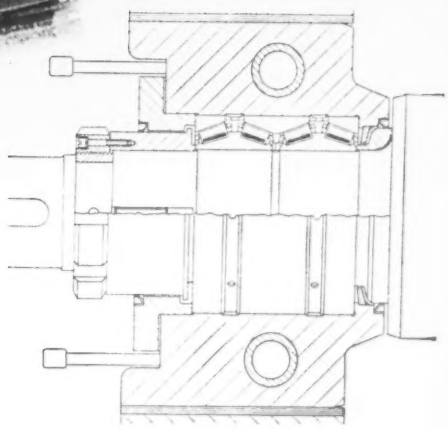


On the five-stand cold-reduction mill at Velindre



This five-stand cold-reduction mill built for The Steel Company of Wales, Velindre, by Davy and United Engineering Company Limited, is equipped with Timken tapered-roller bearings on the work roll necks and in the screw-down mechanism. The illustration shows the mill with spare work roll assemblies in the foreground, ready for a work-roll change-over.

The sectional drawing shows a typical arrangement of Timken tapered-roller bearings on work rolls.

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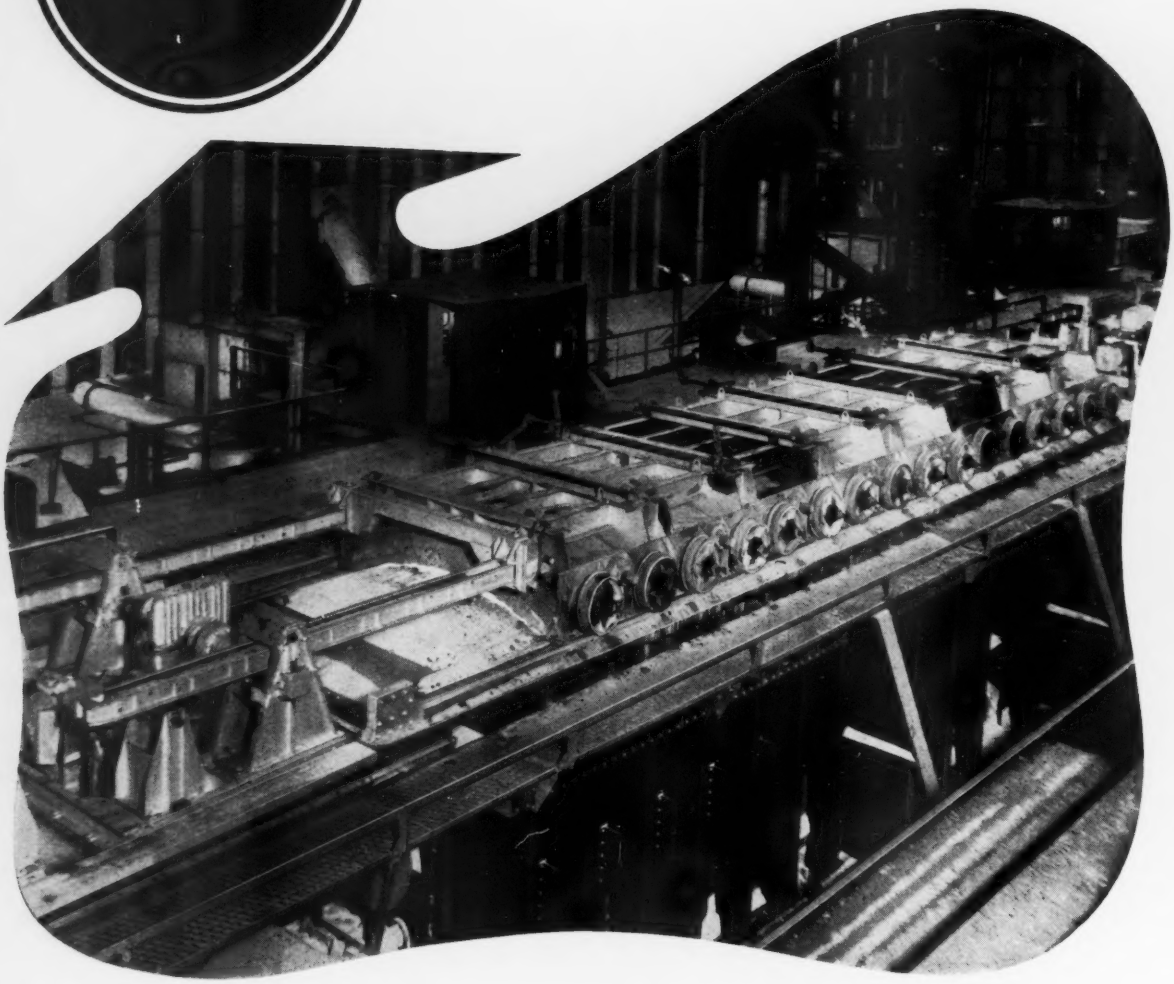

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The above illustrates one of a battery of blast furnace gas fired reversing regenerative Soaking Pits, each with a holding capacity of 120 tons.

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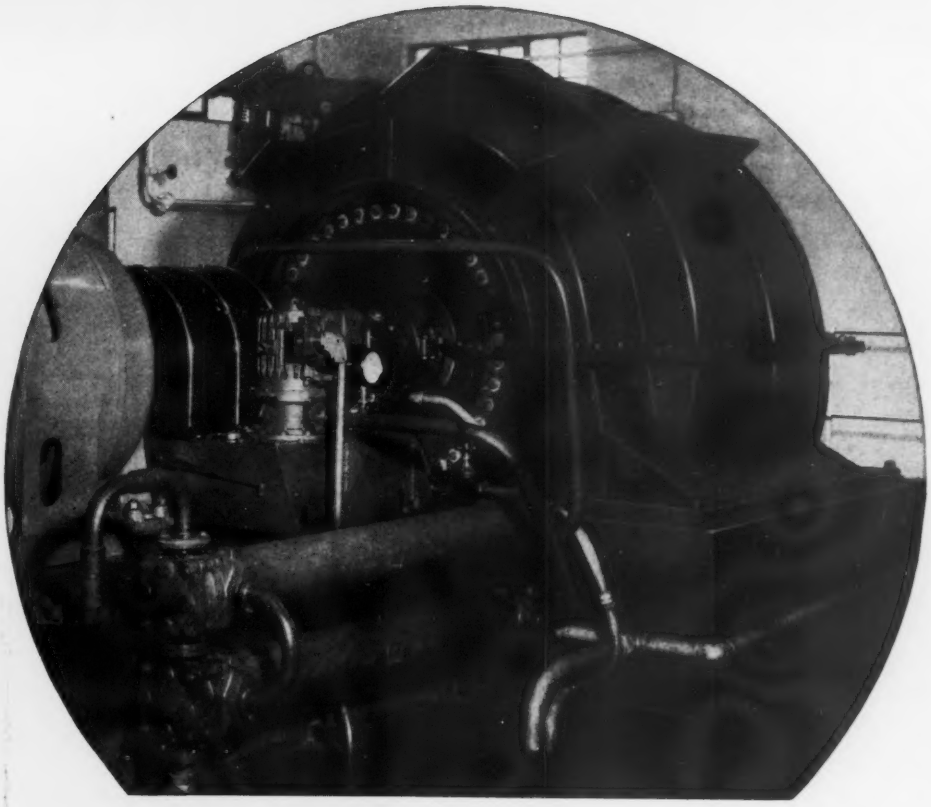

TWO-STAGE GAS BOOSTER

The illustration reproduced by the courtesy of the West Midlands Gas Board shows the Howden two-stage Booster with variable speed drive.

This installation is at Washwood Heath Gas Holder Station, and is designed to handle an inlet volume of 2,250,000 cubic feet per hour of Town's gas at a differential pressure of 80" wg. The total adiabatic efficiency of the Booster is 82%

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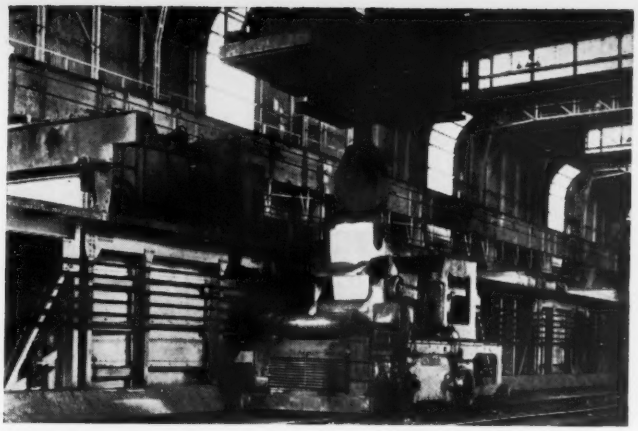
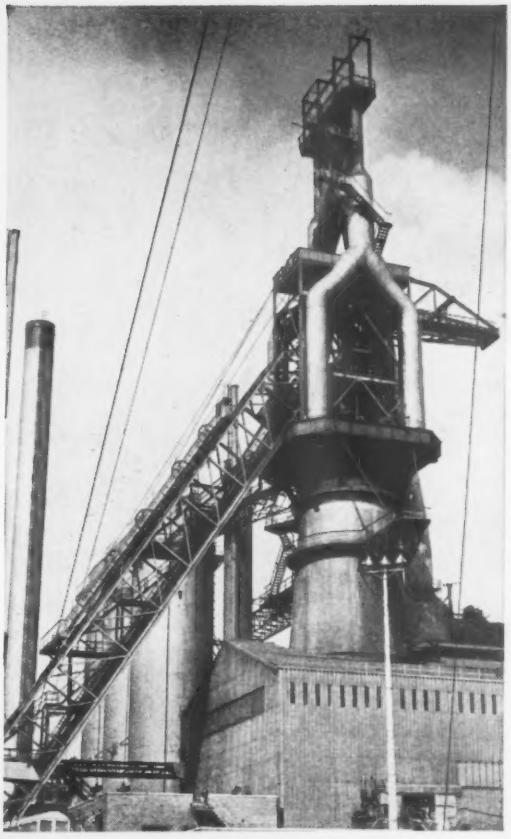
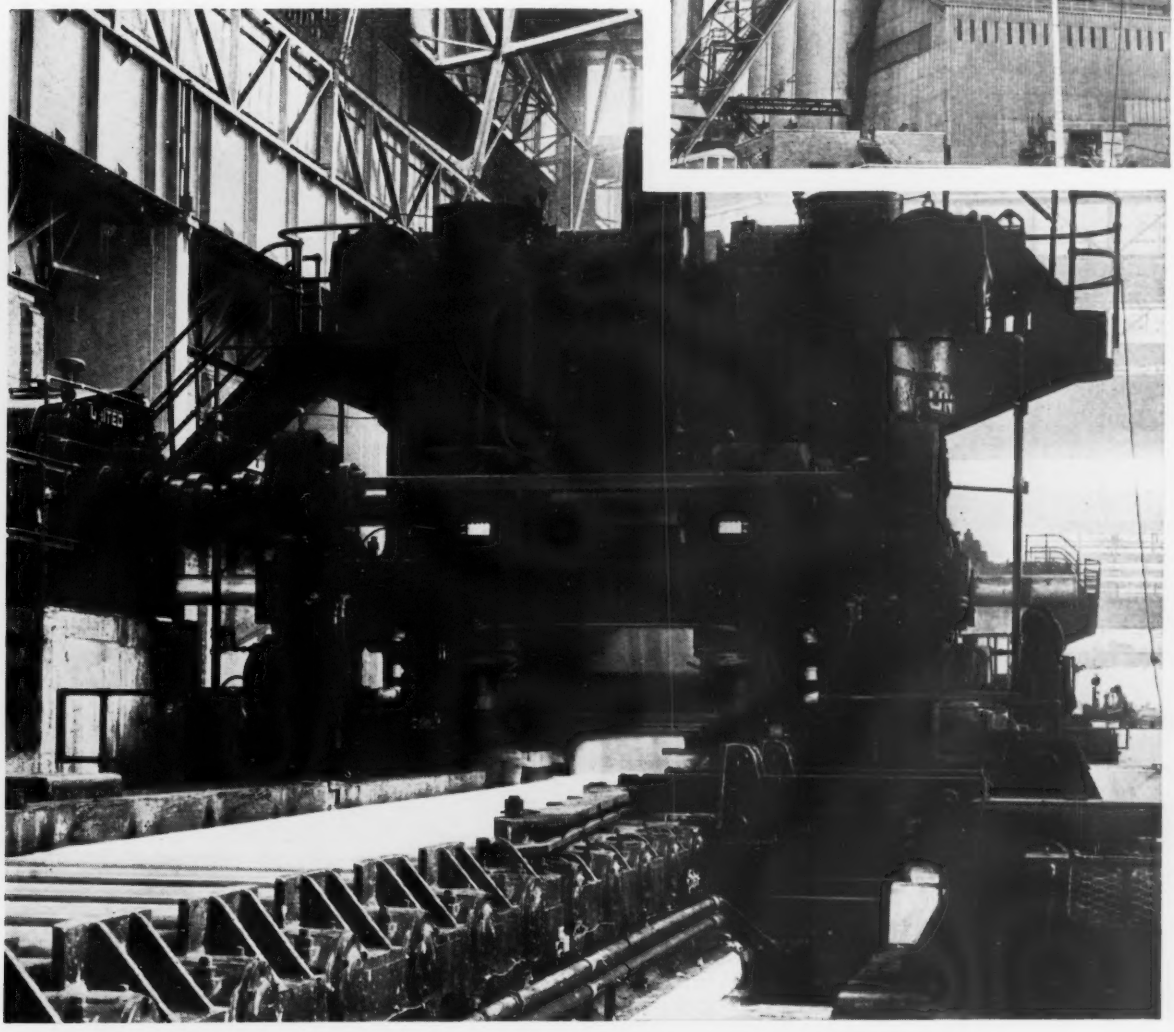


After leaving the soaking pits: an ingot being placed on the ingot buggy



No. 4 Blast Furnace

A slab passing through one of the roughing stands of the 80" continuous hot-strip mill at Abbey Works



At de Havilland a G.E.C. Furnace is used in forming Britannia propeller blades

For efficient process heating use

G.E.C.

The hollow steel blades of a Bristol Britannia propeller are manufactured in two parts—an outer shell, made from sheet steel, and an inner core made from steel tube. After initial cold manipulation, these components are hot-formed to give the required blade shape; for this purpose de Havilland Propellers Ltd. chose a G.E.C. Horizontal Batch Furnace. During heating, close control over the carbon content of the steel is maintained by an atmosphere produced with G.E.C. plant.

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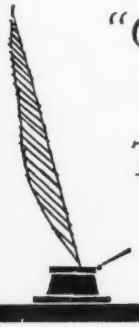

"O! for a muse of fire,
that would ascend
The highest heavens
of invention."

Thus Shakespeare wrote a long time ago, but we feel that it is apt to use his lines in a different context. Since the advent of nuclear power, writers have rivalled each other in sounding the loudest paeans of praise for modern science. Calder Hall indeed stretched their vocabulary and Dounreay will no doubt do so even more. We, as engineers, are more reticent and prefer to say that atomic power is here to stay and leave it at that.

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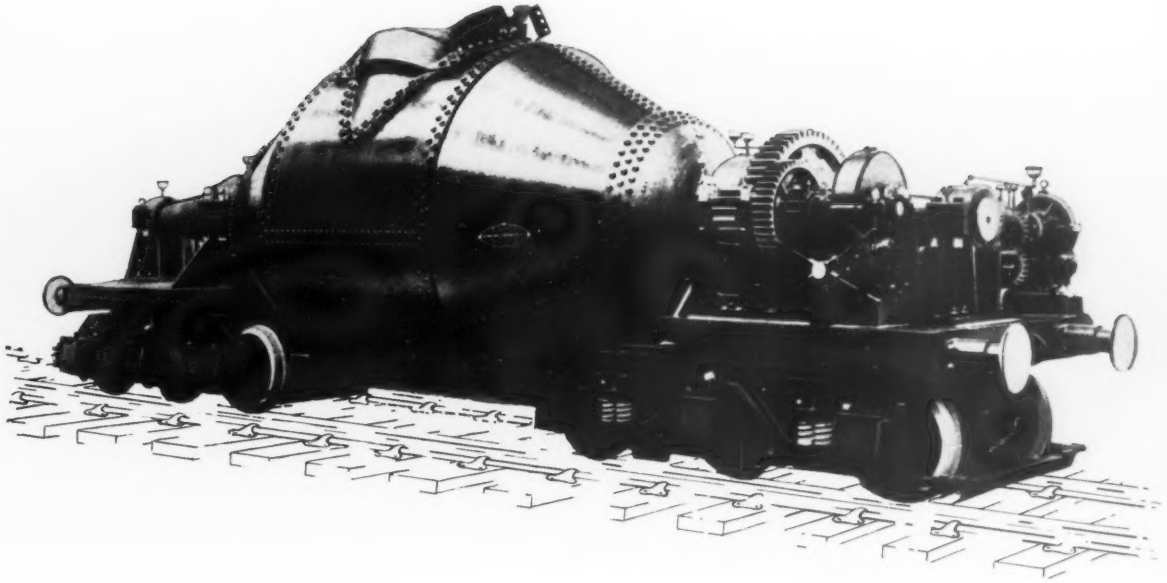

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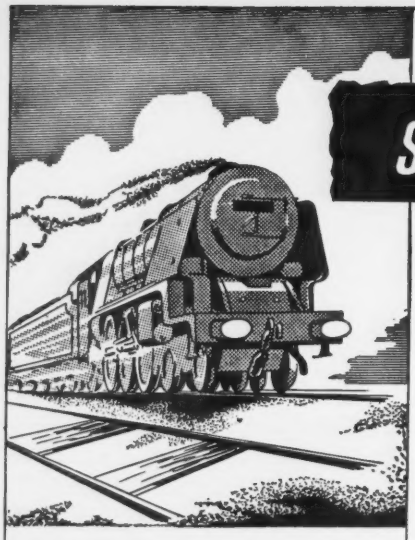

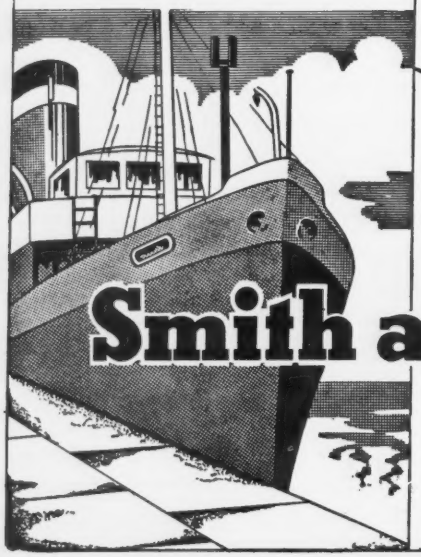

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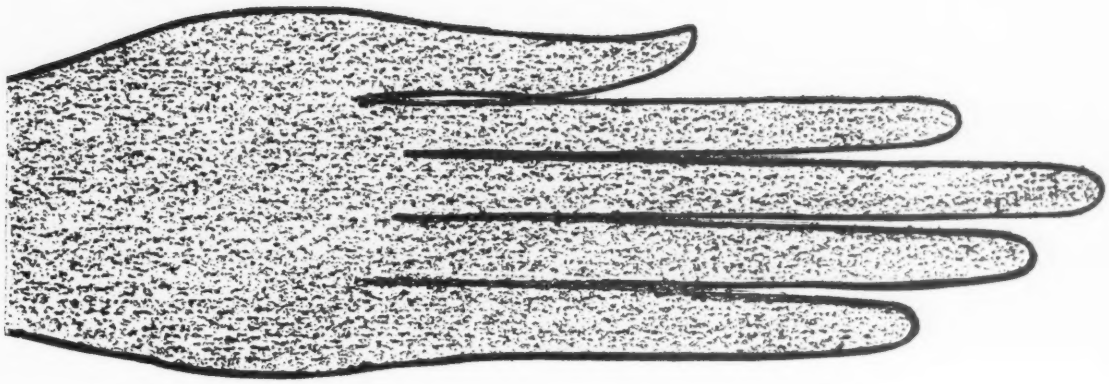

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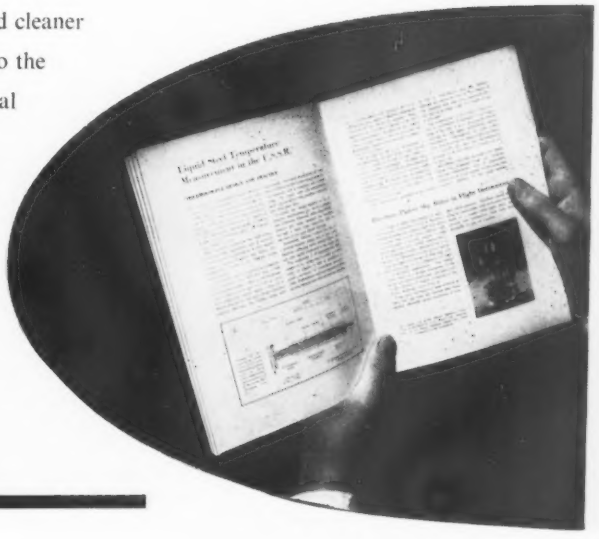

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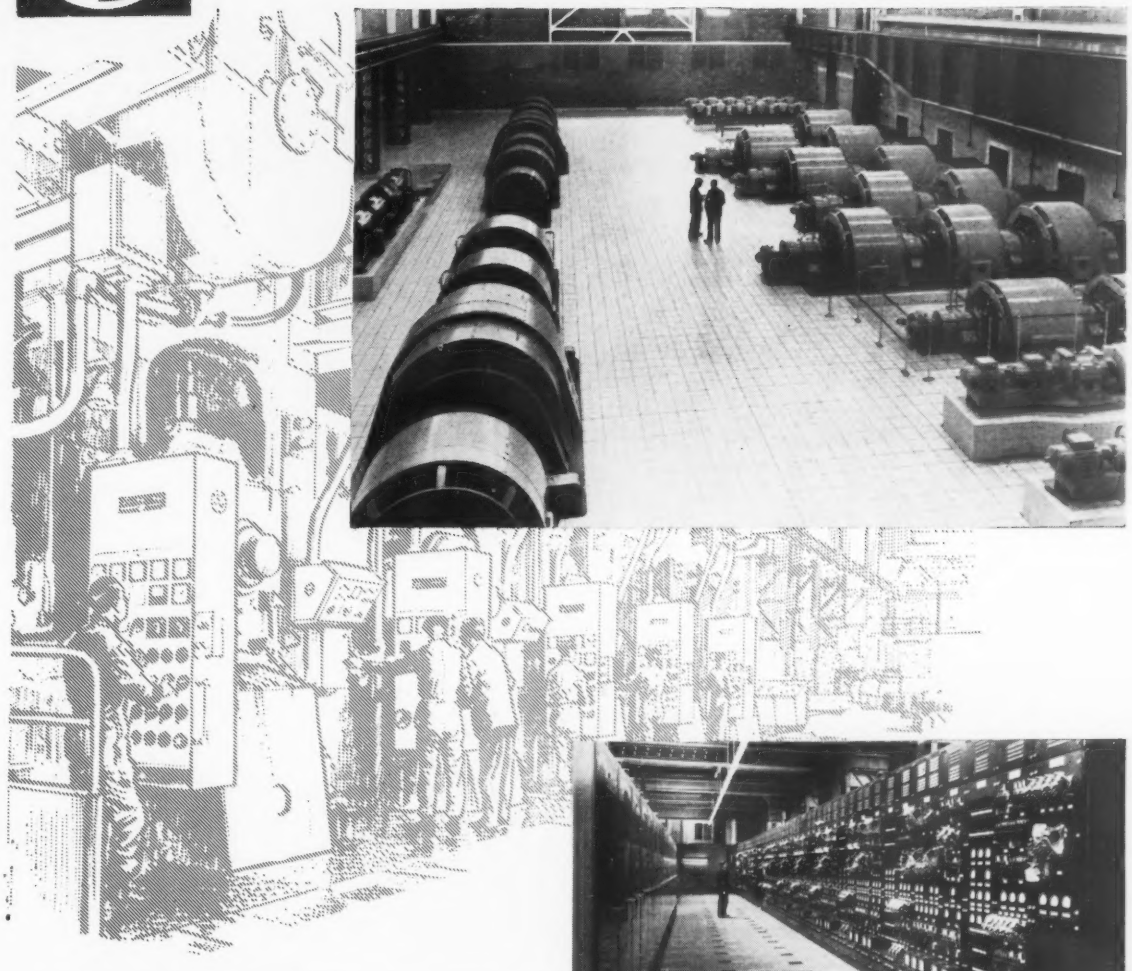

For the Steel Company of Wales' important new plant at Velindre, the main drive for the five-stand tandem cold reduction mill comprises six BTH D.C. motors totalling 19,150 h.p. supplied from two BTH motor-generator sets with a total capacity of 16,730 kW.

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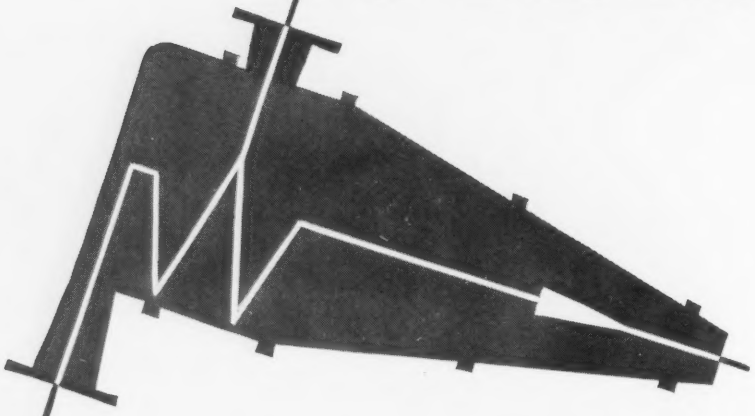

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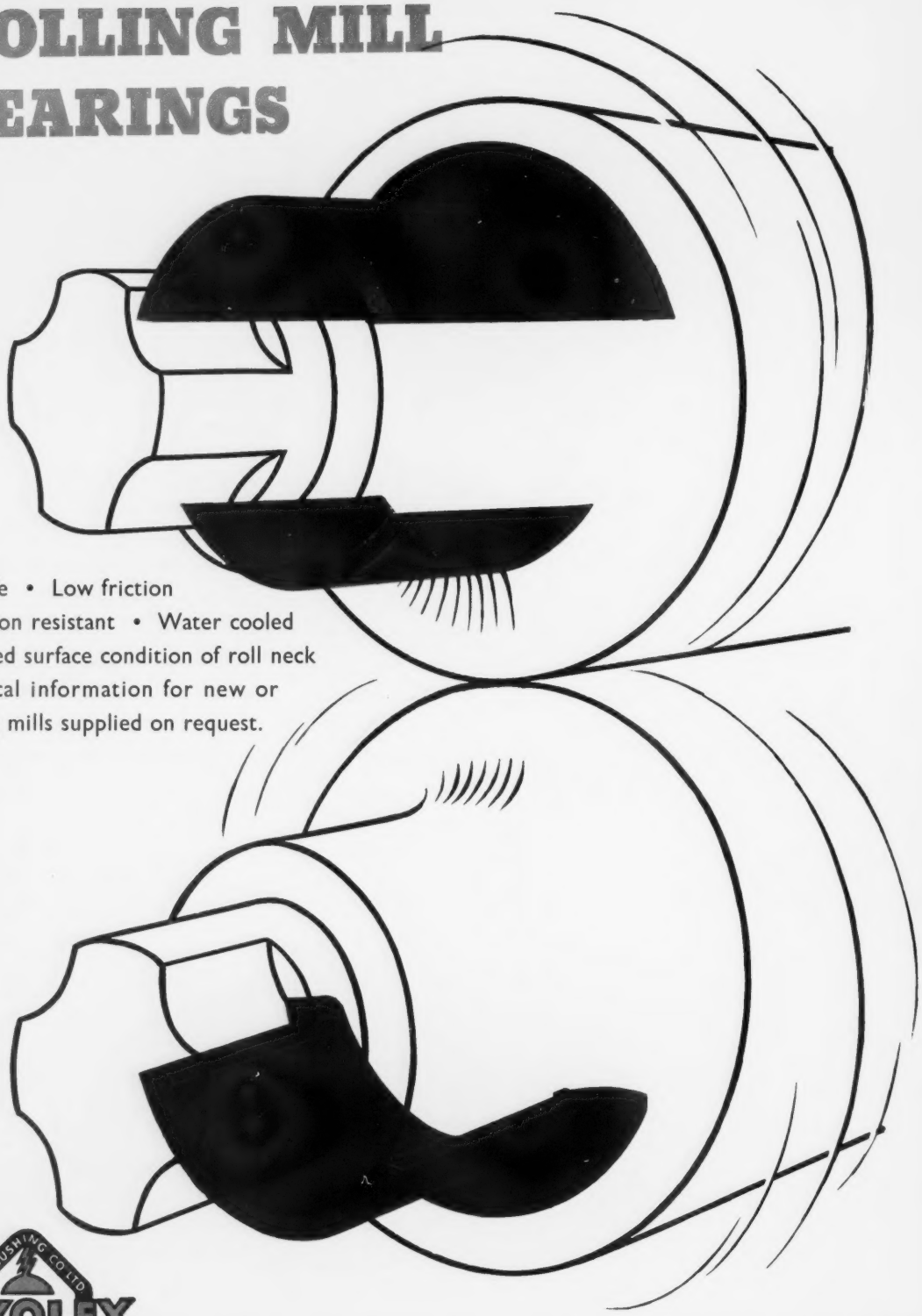

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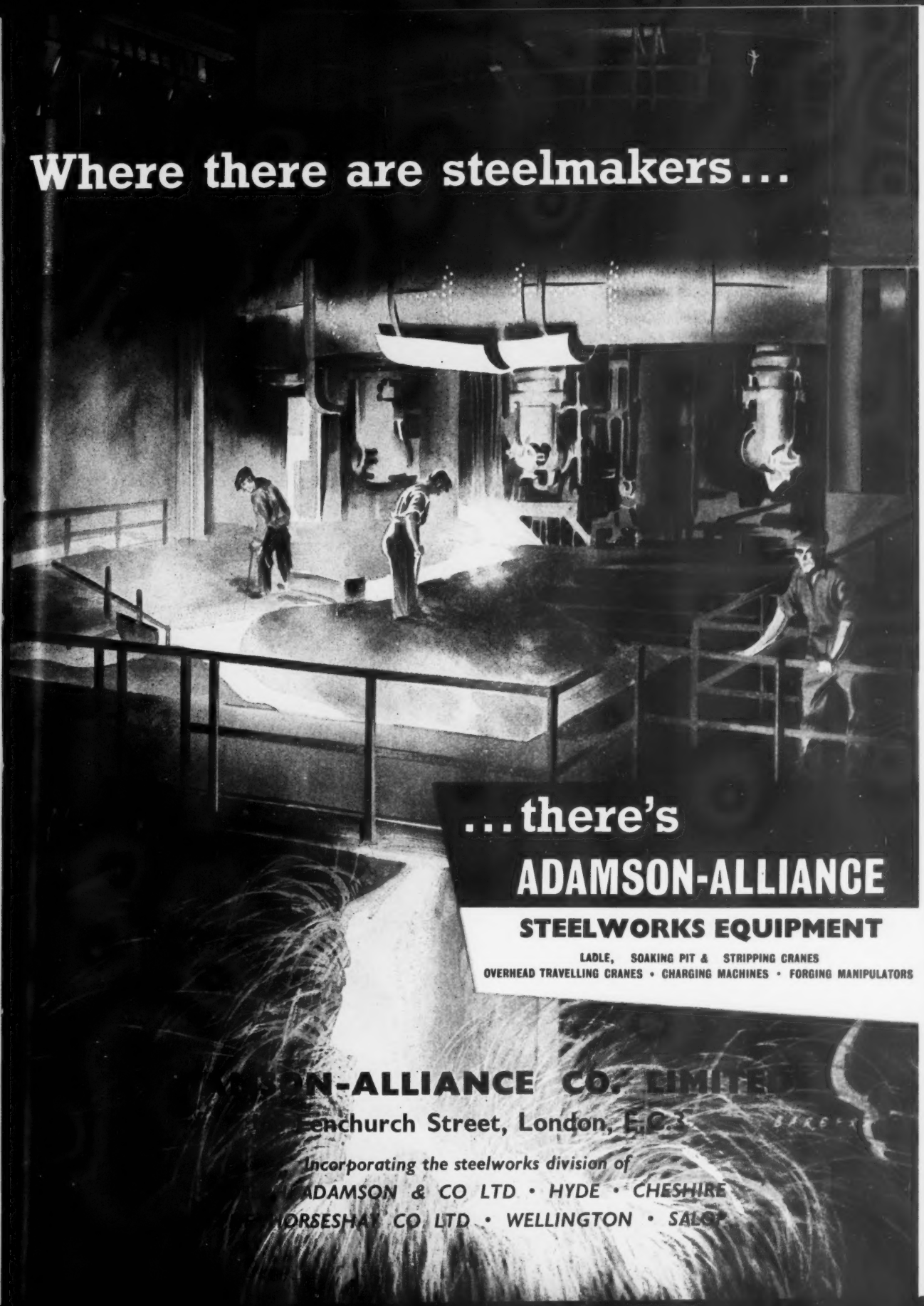

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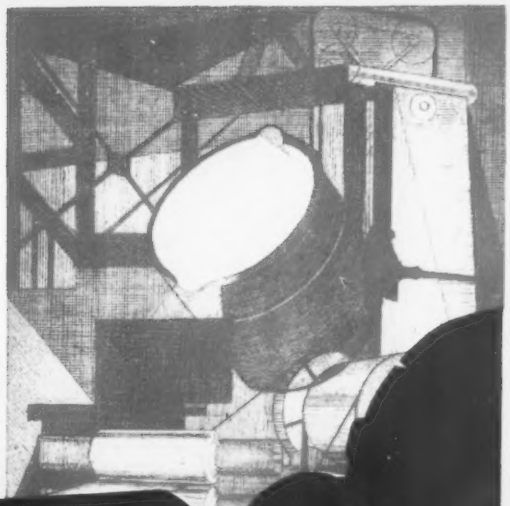
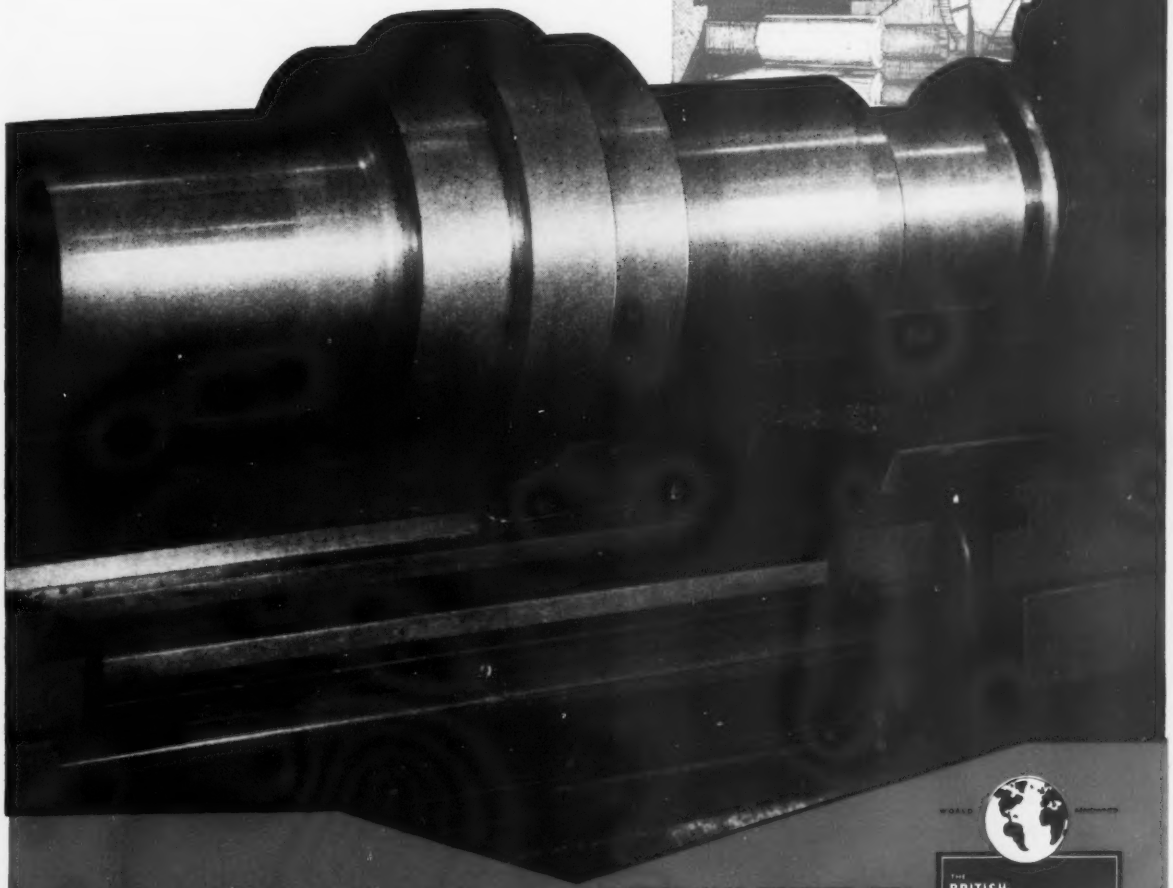


The Symbol of Roll-making (from an original etching)



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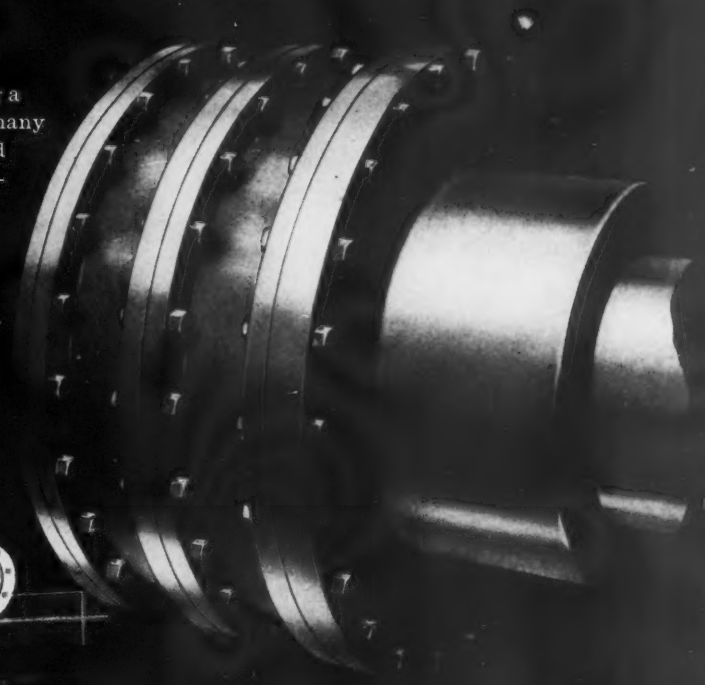

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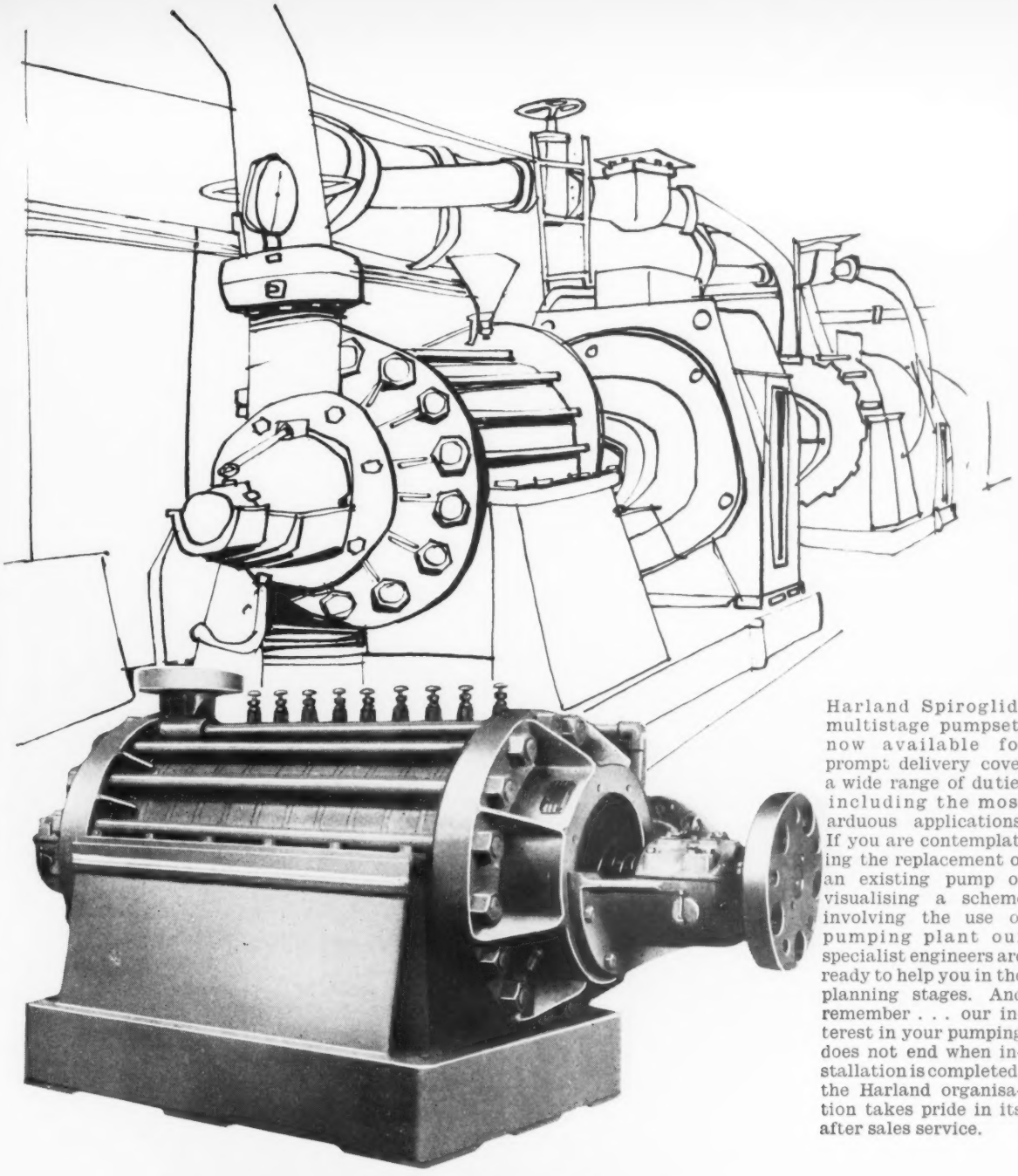

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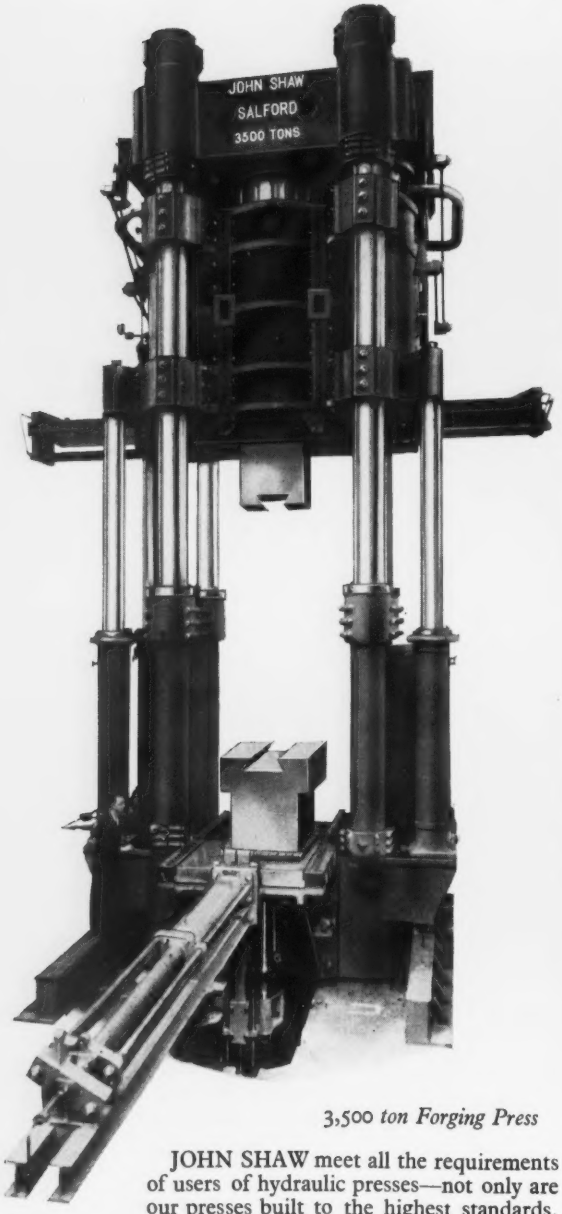

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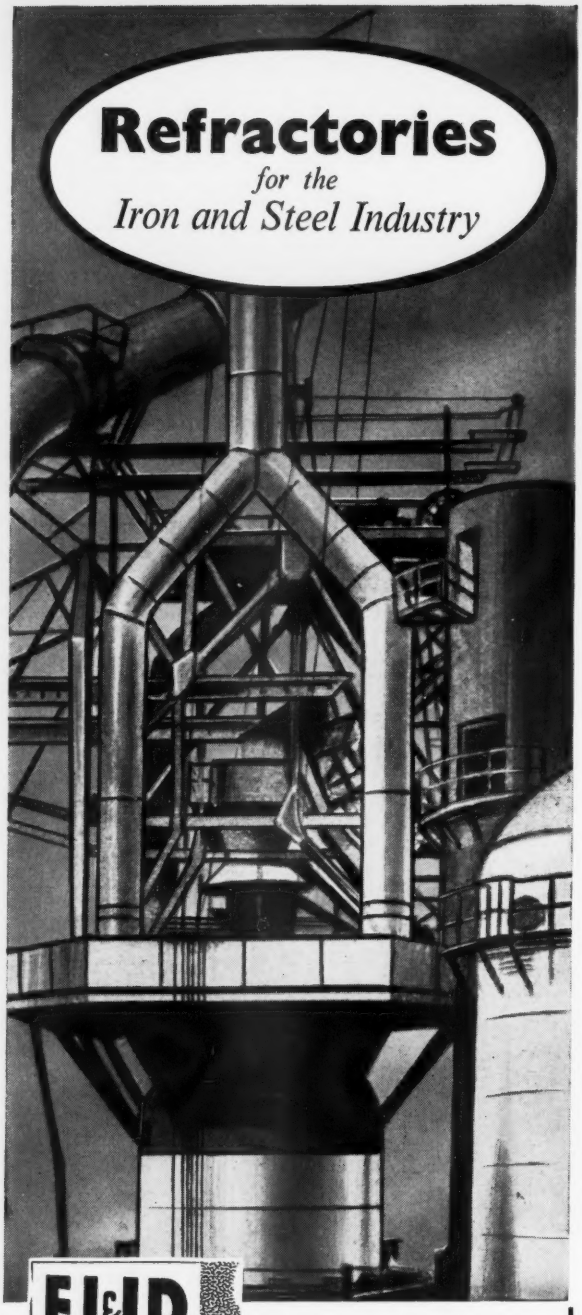

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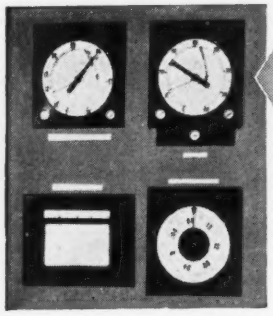

Honeywell Area Flow-meter

Colville's Dalzell Works. Area Meter Body on creosote-pitch line in valve cabin transmits to ElectroniK Inductance Bridge Receiver on control panel, top right.

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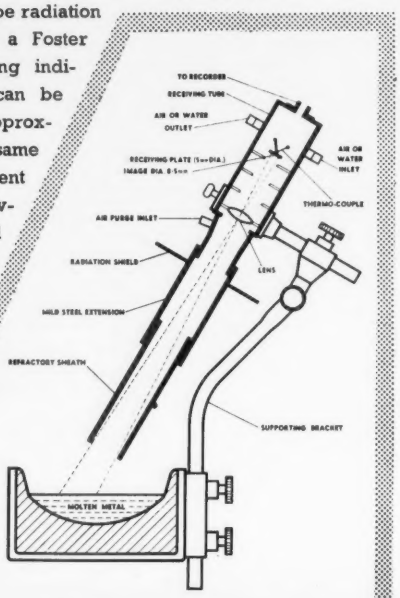

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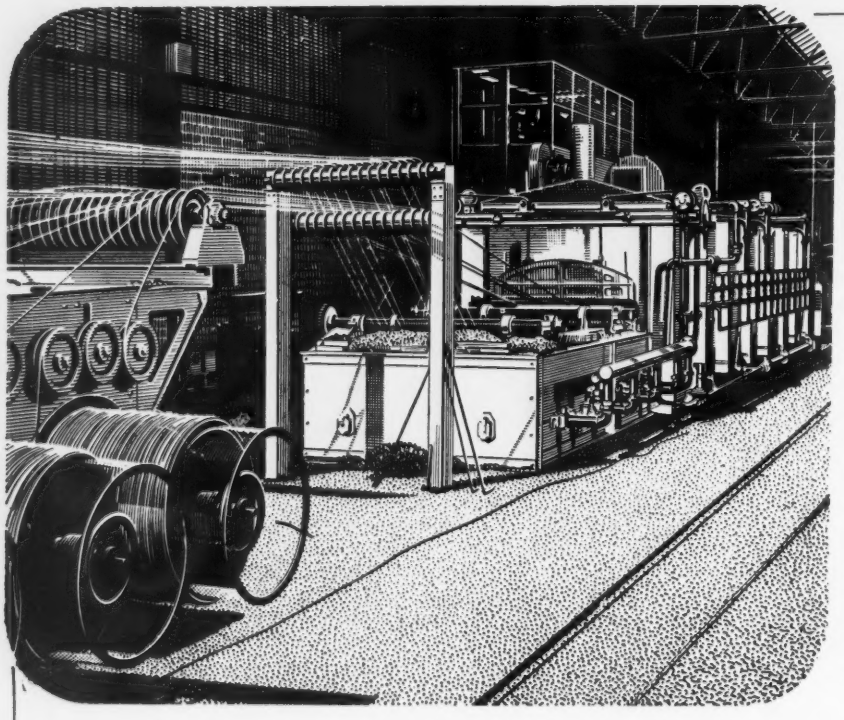

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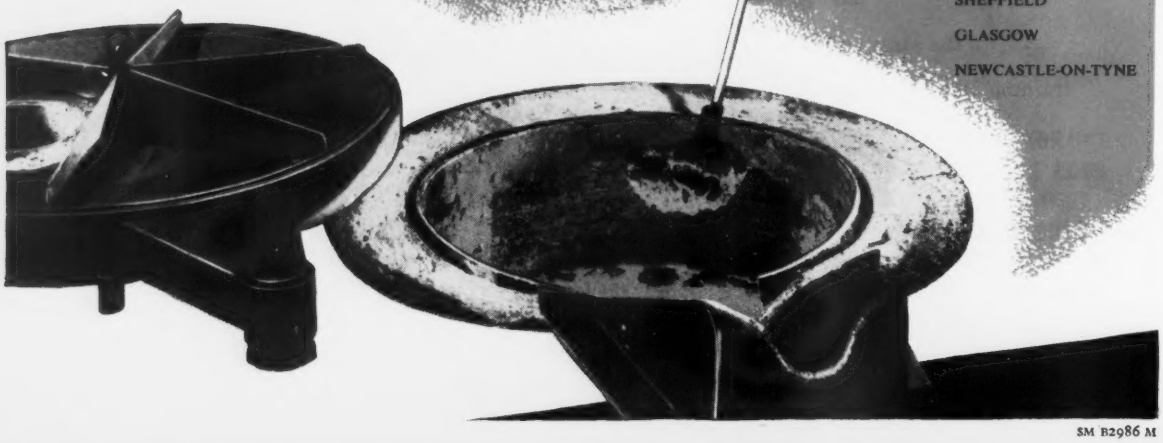

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- Crushes breeze containing up to 16% of moisture.
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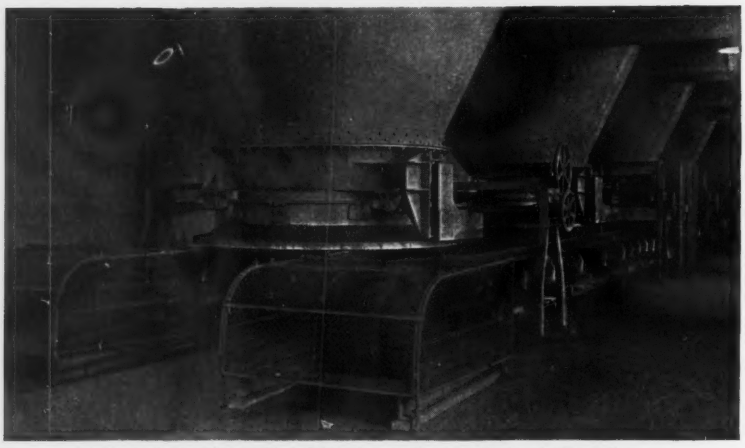

A battery of nine Newells Feed Tables in operation at a Steelworks Sinter Plant

HEAVY DUTY Feed Tables

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- Very reliable in operation.
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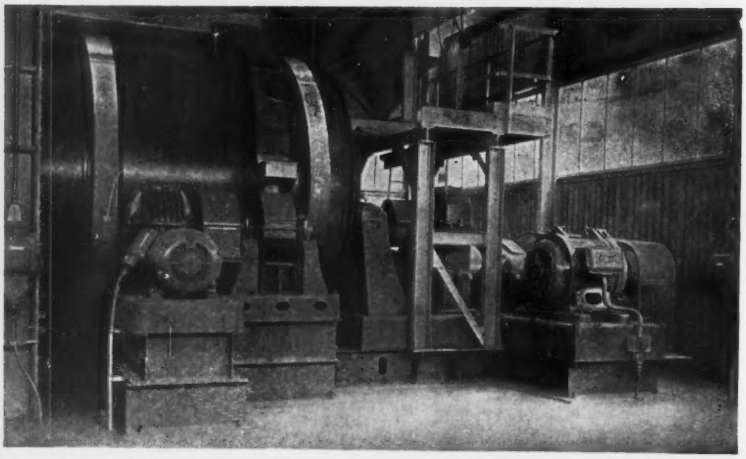

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- Intensive, thorough and uniform mixing of solids and water unobtainable by any other machine.
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Contents

TENTH HATFIELD MEMORIAL LECTURE

The Mechanism of Formation of Banded Structures *by Paul G. Bastien* 281

THE IRON AND STEEL INSTITUTE

Historical Note No. 2 (Second Series)
An Early Invention for Protecting Iron against Corrosion *by H. R. Schubert* 291

Low-carbon Bainitic Steels *by K. J. Irvine and F. B. Pickering* 292

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

Sinter-bed Ignition *by H. Bates* 310

The Tempering of Low-alloy Creep-resistant Steels containing Chromium, Molybdenum, and Vanadium *by E. Smith and J. Nutting* 314

A Study of the Tuyere Combustion Zone *by J. Taylor, G. Lonie, and R. Hay* 330

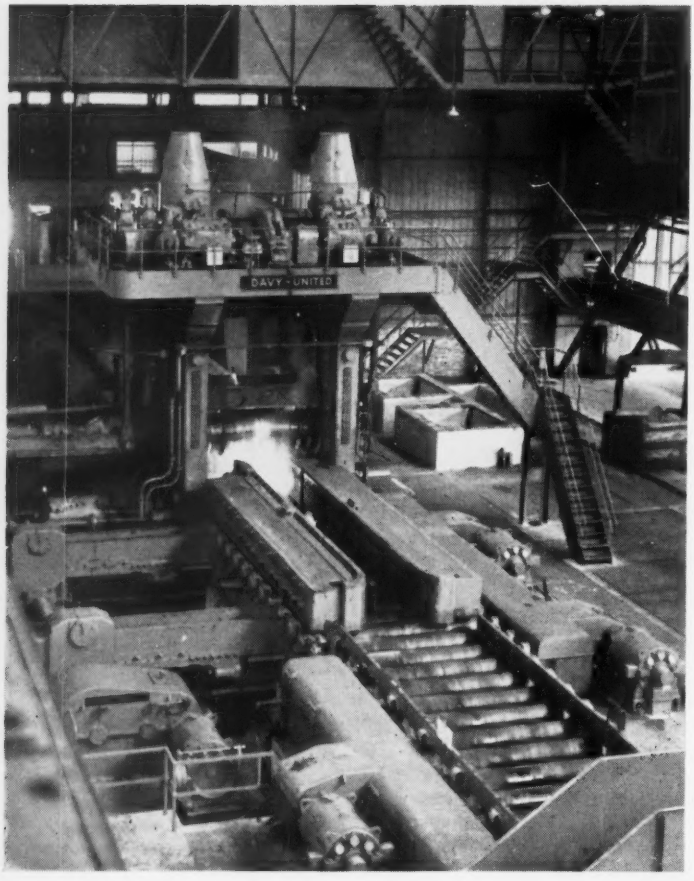
An Investigation of a Method for the Combined Determination of Niobium and Tantalum in Steel *by the Methods of Analysis Committee* 341

NEWS 344 **ABSTRACTS** 349

TRANSLATIONS 348

THE IRON AND STEEL INSTITUTE, 4 GROSVENOR GARDENS, LONDON, S.W.1

ROLLING MILL PLANT
for the continuous production of billets and slabs



From this mill plant at Consett Iron Company Limited come blooms, heavy slabs for plate rolling, and a wide range of billets and small slabs, in total output of about 22,000 tons per week. It consists of a Davy-United 44″ x 108″ High Lift Blooming and Slabbing Mill feeding a 10-Stand Morgan Continuous Billet and Slab Mill, and is planned for straight-line delivery so that progression from ingot to finished product can be carried through without the necessity for reheating. In all, some 7,500 tons of mill machinery were supplied by Davy-United for this contract.

The Blooming and Slabbing Mill rolls ingots up to 20 tons in weight; it is fitted with Morgoil bearings on tapered roll necks, has universal manipulators and employs twin motor drive from two 4,075 h.p. motors.

The Morgan Mill is arranged in two groups of 6 and 4 Stands respectively. This view back down the plant from the 4-Stand Group, shows a Morgan patented electric flying shear, chain transfer from the 6-Stand Group and a Davy 66″ Hot Saw.

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226/D.U.

Professor PAUL G. BASTIEN, D.Sc.

PAUL G. BASTIEN was born in 1907. After attending the *lycées* at Toulouse and Algiers, he entered the Ecole Centrale at Paris, where he graduated in 1929, first in his year, with the Diploma of Ingénieur des Arts et Manufactures. His studies continued under Professor Albert Portevin, and, as a result of researches into ultra-light Mg–Al–Cu alloys and the flowability of metals and alloys, he was awarded the degree of Doctor of Physical Sciences at the Sorbonne. Shortly afterwards he obtained the degree of Doctor of Law.

Professor Bastien has combined teaching with industrial work since the beginning of his career; he has always striven to relate his theoretical studies to his more applied researches and to explain the fundamental causes or mechanisms of the phenomena encountered during his studies in the metal industries (principally in iron and steelmaking).

In the educational field he has been successively lecturer, head of metallographic and X-ray studies, and (since 1942), Professor of General and Metal Physics at the Ecole Centrale des Arts et Manufactures. He succeeded his distinguished teacher, Professor Portevin, as Director of Scientific Research there in 1953. He is also a Professor at the Ecole Supérieure de Fonderie, the Institut de Soudure, and the Centre d'Etudes Supérieures de la Sidérurgie.

Professor Bastien's connection with industry began with the Services Techniques de l'Aéronautique. After some years he joined the Etablissements Schneider as consultant; he was appointed to his present position of Scientific Director in 1946. When the Société des Forges et Ateliers du Creusot was formed in 1950, Professor Bastien became its Scientific Director, a post he still occupies. He is also a member of the scientific and technical committees of the Institut de Recherches de la Sidérurgie (IRSID) and has been a consultant to the Commissariat à l'Energie Atomique since 1949.

Professor Bastien's studies have covered a very wide field, including the Mg–Al–Cu equilibrium diagram, the flowability of metals and alloys, the heterogeneity of steel ingots, physico-chemical reactions in 18/8 and Fe–Cr alloys, and the influence of hydrogen. The work carried under his direction on the last-named subject at the Centre de Recherches de Physique des Métaux at the Ecole Centrale has been of very great importance in the specification of steels which will withstand the attack of wet sulphurous natural gas.

In 1952 Professor Bastien was made a Laureate of the Académie des Sciences, and in 1957 he was appointed an Officier de la Légion d'Honneur. He is a past-president of the Société des Ingénieurs Civils de France, the Association Technique de Fonderie, and the Société des Ingénieurs Soudeurs. He has been a Member of The Iron and Steel Institute since 1936.

On 28th November, Professor Bastien delivered the Tenth Hatfield Memorial Lecture at Sheffield; the subject of his lecture, which appears on pp. 281–291 of this issue, was "The Mechanism of Formation of Banded Structures".

Professor Paul G. Bastien, D.Sc.

Tenth Hatfield Memorial Lecturer, 1957

TENTH HATFIELD MEMORIAL LECTURE

The Mechanism of Formation of Banded Structures

By Paul G. Bastien, D.Sc.

When I received the invitation to give the 10th Hatfield Memorial Lecture, I felt both greatly honoured and at the same time rather apprehensive at seeing my name added to the list of eminent metallurgists who have been called upon to honour the memory of the late Dr. Hatfield by delivering this Lecture.

Some years before the second World War I had the opportunity to meet Dr. Hatfield on several occasions, and I was very greatly impressed by the breadth of his knowledge, his depth of thought, and his method of expression, which was always extremely clear, often even jocular, and always to the point.

The choice of a subject for this Lecture, which should be connected with Dr. Hatfield's work, is a relatively easy task, in view of the range and variety of his activities. For many years he served as the distinguished Chairman of the Heterogeneity of Steel Ingots Committee of The Iron and Steel Institute, and for that reason, and also because the 10th Hatfield Memorial Lecture is being given in Sheffield, a town well known in the world of metallurgy, it seemed appropriate to discuss on this occasion the mechanism of formation of the banded structure in steel, a subject in which I have taken a great interest over many years.

MANUFACTURING PROCESSES which involve plastic deformation without the removal of metal have an important effect on the structure of a metal or alloy, not only on the macro- and microscopic scale but also on the crystalline structure; furthermore, this effect differs greatly, depending on whether the deformation takes place at high or low temperature.[1]

Several phenomena are known to occur during the non-isothermal solidification of an alloy over the solidification range. In the case of steel, they cause the composition to vary in the different parts of the material and give rise to the following effects:

Large-scale segregation, variations in concentration between the liquid and solid phases during solidification which are found over the whole of the ingot. This has formed the subject of a large number of investigations, among them the outstanding studies of the Heterogeneity of Steel Ingots Committee of The Iron and Steel Institute.[2]

Small-scale or dendritic segregation, which affects the crystals of primary crystallization and implies the existence of solid solutions in which the concentration varies with temperature.

In a worked steel we have to consider the following structures resulting from the plastic deformation:

A macroscopic laminated structure caused by dendritic segregation

A microscopic laminated structure, which is particularly marked if forging takes place between the A_3 and A_1 points or below the A_1 point

A crystalline laminated structure, which results from the actual mechanism of the plastic deformation, mainly slipping.

One of these structures will predominate and control the properties of the final product, depending on the type of steel and its production, the method of casting, the weight and shape of the ingot, the actual conditions under which the plastic deformation takes place, in particular the range of temperatures and the amount of deformation, and finally the intermediate reheats and the final heat-treatment.

This lecture was delivered at the Firth Hall, Sheffield University, on 28th November, 1957.

Fig. 1—Laminated structure of forging. Etched in Stead's reagent × 1

THE BANDED STRUCTURE: DESCRIPTION AND CONDITIONS FOR ITS APPEARANCE

During the solidification of steel, the most impure liquid with the lowest melting point, which contains in particular sulphur inclusions in the form of suspensions, becomes concentrated and trapped in the spaces within the dendritic network.

It has been suggested, first by Stead[3] and later by Howe,[4] Sauveur,[5] and Whiteley,[6] that during the cooling through the transformation range the phosphorus concentrated in the interdendritic spaces displaces the carbon from these areas towards the arms of the primary crystals. According to this view, the process results in the presence of dendrites whose arms are rich in carbon and poor in phosphorus, while the interdendritic spaces have a low carbon and a high phosphorus content.

This effect, however, appears only in steels with a relatively high phosphorus content and for suitable rates of cooling through the transformation region. Several years ago we had the opportunity to examine a number of 300-mm (12-in.) dia. forged cylinders after quenching and annealing, made from Ni–Cr steels produced in 6-ton ingots in acid, basic, and duplex electric furnaces.[7] These steels had a low sulphur and phosphorus content (S 0·011–0·031%; P 0·015–0·021%) and a suitable cooling rate (water quench after a homogenization reheat). It was found that because of the low phosphorus content there was no displacement of carbon towards the dendrite axes, the carbon being concentrated in the original interdendritic spaces in accordance with the usual effect of small-scale segregation.[8]

During solidification, the inclusions themselves are also subject to large and small-scale segregation. The silicate inclusions are concentrated in the arms of the dendrites, giving a certain amount of support to Benedicks's theory[9] that inclusions rich in oxygen act as nuclei for crystallization from the liquid during the solidification of the ingot. The sulphur inclusions, on the other hand, are located in the interdendritic spaces in accordance with the usual mechanism of small-scale segregation.

This small-scale segregation would disappear if the inclusions could be absorbed by means of a long heat-treatment at the high temperatures of the gamma region, thus making the material entirely homogeneous chemically. This is never the case in practice, mainly because of the low diffusion rates of phosphorus and most other dissolved elements (except carbon), and also because of the very low solubility of non-metallic inclusions in the solid state in iron. This segregation therefore persists, although to a more or less reduced extent, even after the high-temperature homogenization heat-treatments used for large forgings.

To simplify matters, let us consider an ingot hot-worked by forging or rolling, after complete solidification but without return to ambient temperature. The plastic deformation has the effect of modifying the

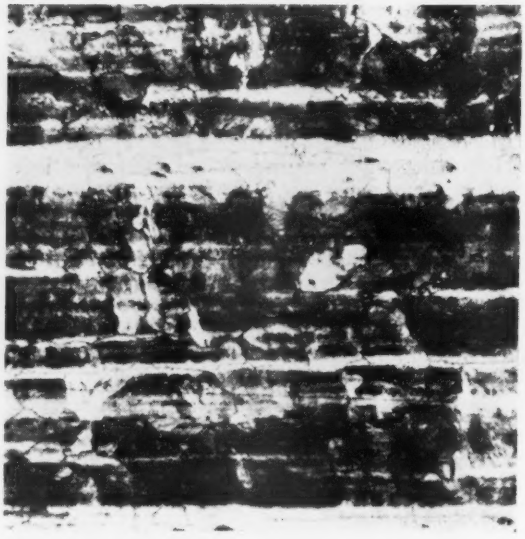


Fig. 2—Macroscopic banded structure. Etched in Stead's reagent × 100

dendrites by aligning them and deforming them mainly in the direction of the metal's greatest elongation; it follows that the inclusions must tend to form layers or laminæ. This results in a fibrous or laminated structure which, in the case of rolling or forging, can readily be demonstrated by the use of metallographic reagents, such as cupric chloride or iodine (Fig. 1).

During cooling, the separation of the proeutectoid constituents depends on the local chemical composition and is therefore initiated by the dendritic segregations which have been arranged in an orderly manner during hot working. When there is a pronounced phosphorus segregation, a micrograph shows a banded structure in the deformed material (Fig. 2); the white bands correspond to interdendritic regions which, as shown above, are in general rich in phosphorus and poor in carbon and have a relatively high inclusion density, particularly manganese sulphide. The presence of manganese sulphide inclusions and the phosphorus segregation appear to be related phenomena.

A high-temperature reheat followed by rapid cooling reduces the intensity of the banded structure without suppressing it entirely. If cooling takes place slowly, the banded structure reappears and can be even more pronounced than before.[42] This is because the phosphorus segregation is only reduced very slowly at high temperature and, if we accept Stead's theory for the moment, it continues to displace carbon from the bands with a high phosphorus content, an effect which becomes more pronounced as the rate of cooling through the Ar_3–Ar_1 transformation range is reduced. For that reason the effectiveness of a homogenization treatment diminishes with increasing size of the forging, and especially near its centre.

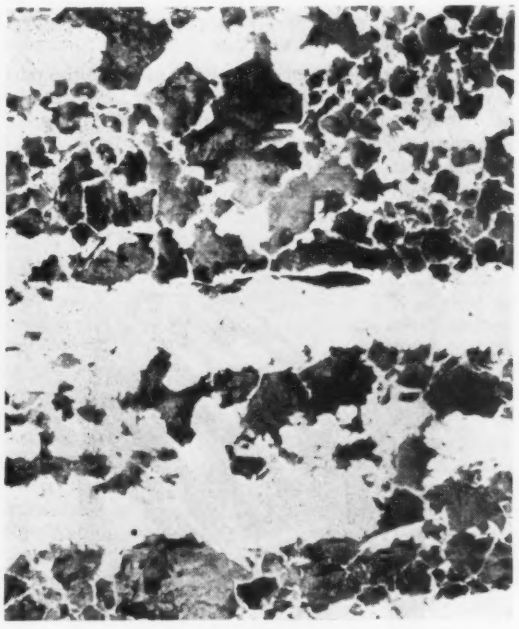


Fig. 3—Banded structure with no carbon displacement; silicates between bands of ferrite × 100

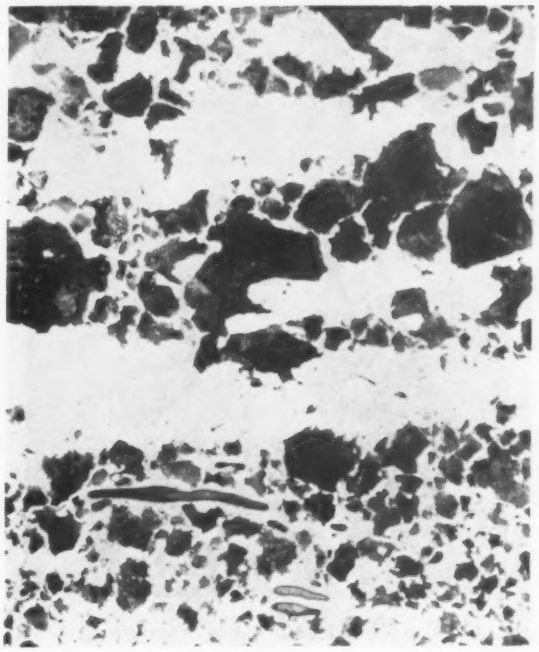


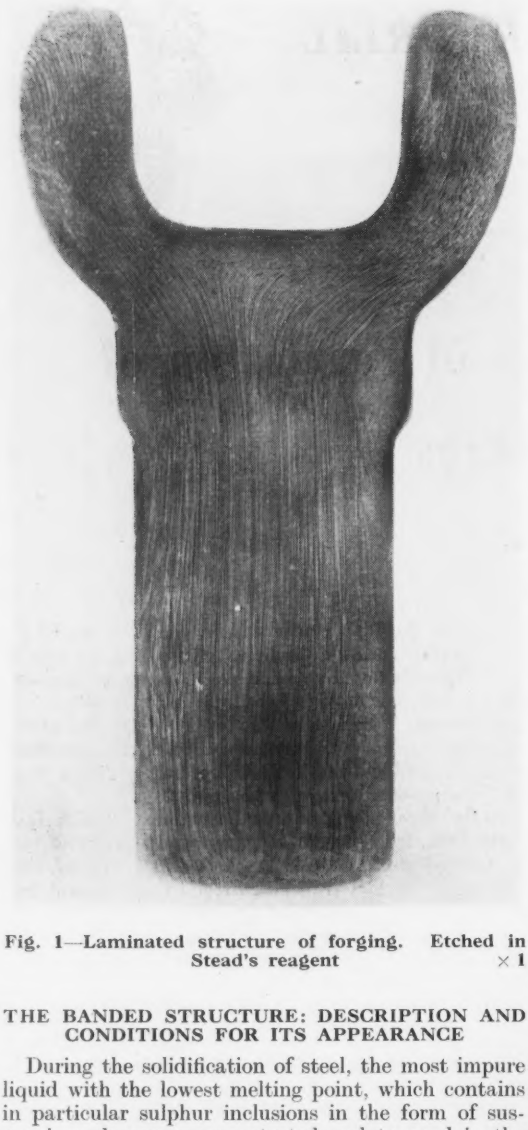
Fig. 4—Banded structure with no carbon displacement; sulphur in carbon-rich areas, characterized by bands of pearlite–ferrite aggregates × 100

The segregation of ferrite in the case of hypoeutectoid steels (or of cementite in hypereutectoid steels) thus results from an excessively slow cooling through the Ar_3–Ar_1 range in the presence of a pronounced phosphorus segregation. The accumulation of inclusions in the interdendritic regions can perhaps also contribute towards the effect to some extent, with the interfaces between the inclusions and the austenite acting as starting points for the growth of proeutectoid crystals.

During the investigation into the Ni–Cr steels with low phosphorus content, in which there was no carbon displacement towards the dendrite axes, we found that the banded structure can also appear in the course of heat-treatment. The cooling rate of the forging during quenching decreases continuously from the skin towards the centre; whenever it has a suitable value at the instant when the transformations take place, the small-scale segregation produces structural heterogeneity by the formation of bands which differ micrographically and which extend along the fibre of the forging (Figs. 3 and 4). Structures characteristic of annealing, quenching, and tempering, as well as intermediate structures, can occur together in the same microscopic field (Fig. 5).

This simple result is very important and suggests already that phosphorus is not the only cause of banded structures but that all the elements which contribute to the production and growth of small-scale segregation must influence this phenomenon.

In the case of the Ni–Cr steels, the sulphur inclusions are always found in those regions which are the



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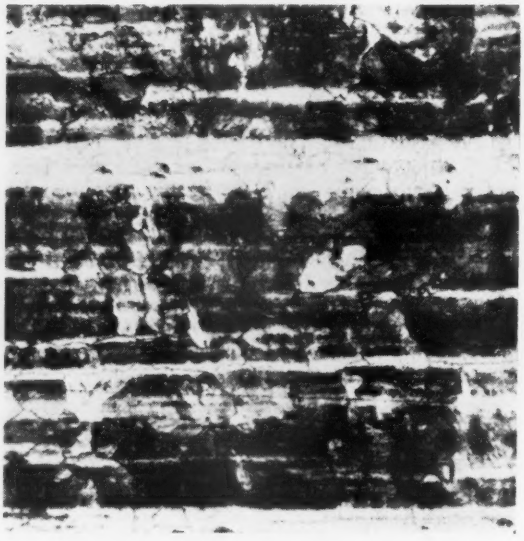

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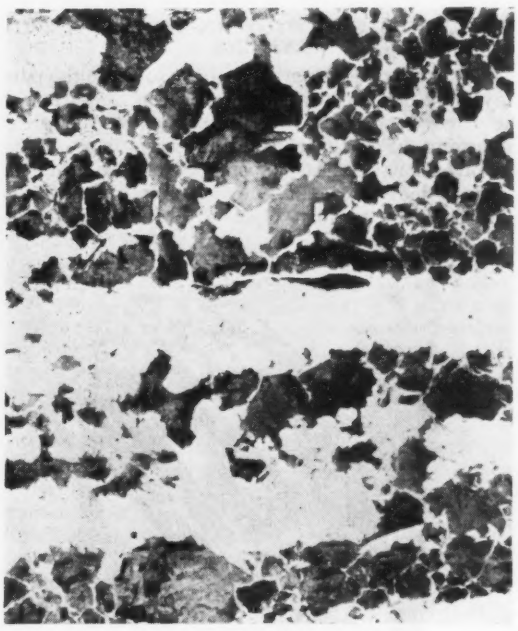

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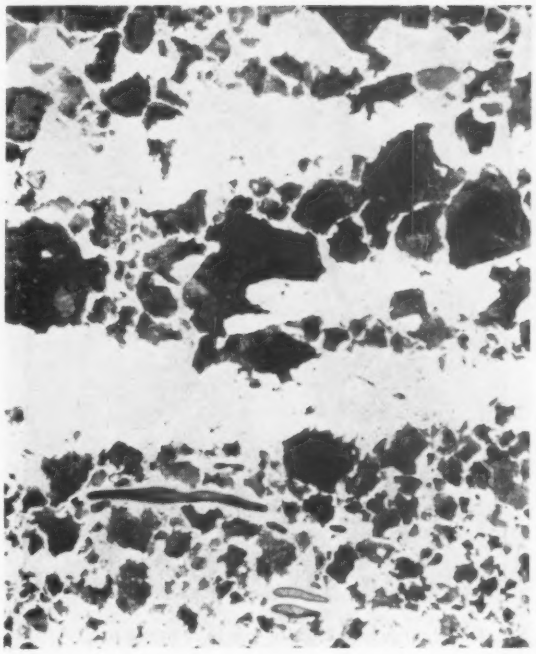

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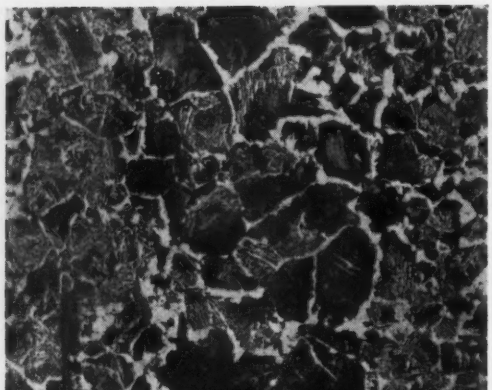

Fig. 5—Pearlite and bainite bands in same microscopic field × 150

most heavily quenched, because they are richest in carbon, and, in the absence of carbon displacement by phosphorus, correspond to the original interdendritic spaces during solidification (Fig. 4). The silicates, on the other hand, are located in the bands of the annealed structure, within the arms of the original dendrites, and more particularly in the ferrite areas (Fig. 3).

To summarize, it is evident that a macroscopic laminated structure is formed through the collapse of a heterogeneous dendritic structure even when forging entirely above the A_3 point. During subsequent heat-treatment this gives rise to a banded structure consisting of layers of similar microscopic constituents with equiaxed crystals, which can readily be detected under the metallurgical microscope. The shape and mechanism of formation of these structures varies, depending on the chemical composition, the impurities, and—at least so it appears at first—on the phosphorus content and the cooling rate at the instant of the transformations in the area concerned.

Let us now consider a forging worked below the Ar_3 temperature. If we take the case of a hypoeutectoid steel, a deposit of proeutectoid ferrite is formed from the austenite as the temperature falls, and the plastic deformation contributes to the formation of laminæ of these two constituents; during the passage through the Ar_1 point the austenite is converted to pearlite. If there is no recrystallization of one of the constituents, pearlite, or of both pearlite and ferrite, the microstructure remains permanently deformed and shows laminæ or bands of constituents of the same type which extend in the direction of drawing and are in some cases markedly non-equiaxed (Fig. 6). In this way we obtain a different type of banded structure; it is only noted in passing and will not be dealt with in detail since it does not form the subject of this Lecture.

The banded structure proper which is found in worked materials under certain conditions has been investigated especially for annealed hypoeutectoid carbon steels.[10] It is particularly frequent in steels containing between 0·10% and 0·35% C and appears only rarely in steels with more than 0·6% C. It is found more often in fine-grained steels and it is favoured by austenitizing slightly above the Ac_3 point or, better still, between Ac_3 and Ac_1, while a high-temperature austenitizing treatment is sufficient to suppress it in certain cases.

Since the banded structure disappears when the cooling rate exceeds a certain value, it has been proposed[11] to define the susceptibility of a particular steel to this type of structure by this ' critical cooling rate for disappearance ' which varies from one steel to another.

The disappearance of the banded structure is only temporary, as we have seen, since a further heat-treatment followed by slower cooling causes it to reappear. Only a high-temperature treatment over an extended period can permanently reduce to some degree the tendency of a steel to form a banded structure by reducing the dendritic segregation through diffusion. Numerous investigations on this subject[12–15] have shown that this result is very difficult to achieve and that it is necessary in some cases to raise the temperature to 1300° C which, of course, raises other difficulties.

Finally, certain workers[10] have noted a relationship between the width of the bands and the temperature necessary to suppress them (Fig. 7).

DENDRITIC SEGREGATION IN STEEL: INVESTIGATIONS

The considerations outlined above indicate the part played by dendritic segregation in the formation of the banded structure and clearly show the great importance of an experimental investigation.

Conventional methods, using Baumann prints or cupric or iodine reagents, provide only a qualitative picture. They make it possible to demonstrate the

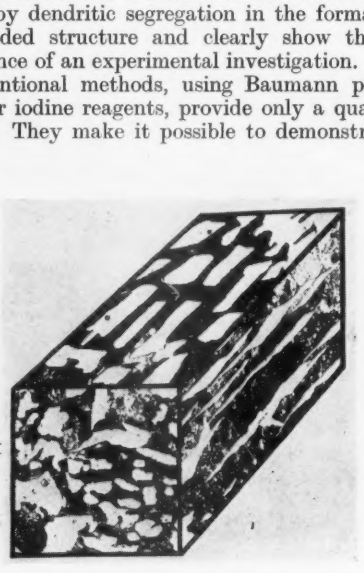

Fig. 6—Cold-worked 0·5% C steel (after Krivobok)

presence of segregation without indicating the nature of the elements involved or the extent of the phenomenon.

Although chemical analyses can readily be used, they involve the development of methods for removing extremely small samples which have to be accurately located in relation to the dendrites. Finniston and Fearnehough[16] carried out a number of microchemical analyses, using absorptiometric methods, on samples removed by means of a $\frac{1}{16}$-in. (1·6-mm) dia. drill. During an investigation at Le Creusot into the heterogeneity of a forged ingot, in collaboration with Cattier, Dubois, and Bleton,[8] we have been able to reveal the different segregated bands by a deep selective acid etch which produced a series of ridges and hollows on the surface of the worked metal. By careful machining it was possible to obtain small quantities of material from the ridges which in this case corresponded to areas richest in the segregated elements. The operation can be repeated ten or fifteen times and produces a sufficiently large sample for the determination of several elements by conventional methods.

The technique of spectrographic point analysis, as developed by Convey and Oldfield,[17] was also used by Finniston and Fearnehough[16] for the investigation of small-scale segregation; several metallic elements could thus be determined in a very small sample of the order of 5×10^{-6} g.

A number of novel methods which are likely to find extensive use in the study of segregations are:

Microradiography
Quantitative autoradiography
X-ray fluorescence point spectrography.

The last of these is of particular importance.

Microradiography

This method, which represents the use of X-ray radiography on the microscopic scale, employs a characteristic radiation K_a chosen so that the various constituents to be investigated differ widely in their absorption coefficients for this radiation. The method has been employed mainly by Trillat,[18] Homès and Gouzou,[19] and Wolfe and Robinson[20]; its first use for the study of segregations in steel appears to have been by Betteridge and Sharpe.[21]

Quantitative Autoradiography

The element whose segregation is to be studied must previously have been activated in a neutron beam from a nuclear reactor. A suitable sensitive emulsion is then applied with certain precautions to the polished surface of the steel sample and the radiation emitted by the labelled element produces a so-called autoradiographic image which indicates its distribution in the surface layer. It is possible to obtain some quantitative information on the distribution of the radioactive element by using a microphotometer and comparing the photographic density at various points of the picture of the sample with that taken from a standard specimen in which the labelled element has a known activity. Using this method Kohn[22] investigated at IRSID the dendritic segregation of phosphorus and arsenic in steel, as well as its persistence during thermal cycles, and developed a special microphotometer for this purpose.

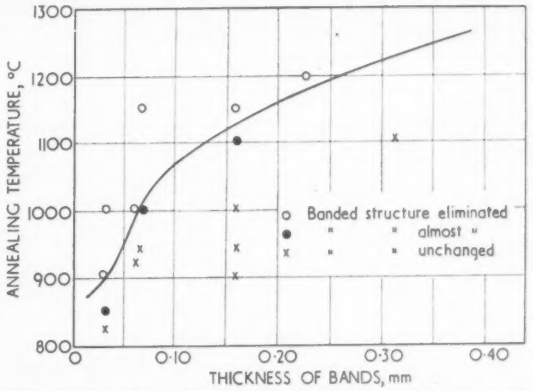

Fig. 7—Relationship between thickness of bands and reheating temperature (after Thompson and Willows[10])

X-ray Fluorescence Point Spectrography

This method was developed by Castaing[23] at the Office National d'Etudes et de Recherches Aéronautiques (ONERA). It consists of projecting a very fine beam of electrons on to the sample and analysing the radiation emitted by the metal after the impact of the electrons, using an X-ray spectrometer. For any given element the intensity of a characteristic radiation emitted by the alloy can then be compared with the intensity of the same radiation emitted by the pure element, so that it is possible to determine the concentration of this element in the volume irradiated; in Castaing's electronic micro-analyser this volume corresponds to a cylinder 2 μ in diameter and 1–2 μ high. All elements whose atomic number is greater than 17 can thus be determined with an accuracy of 1% for concentrations greater than 0·5%.

An equipment of this type is being operated at IRSID and is used for investigations into segregates.[24]

Some of the results obtained in the course of an unpublished investigation into dendritic segregation in a large forging ingot, carried out jointly by IRSID and the laboratory of Le Creusot, are given below.

It is not the intention here to cover the entire literature on the tendency of the various elements in steel to produce dendritic segregation. This would have to include experimental results which cannot readily be compared because the information published depends on

(*a*) *The steel:* chemical composition and impurity content, method of manufacture, casting temperature and weight of ingot which determine the size of the dendrites in the primary crystallization
(*b*) *Its treatment*: working and reheating cycles which influence diffusion
(*c*) *The method of defining segregation:* volume of sample removed, method of analysis
(*d*) *The method of defining intensity of segregation:* certain authors compare the proportions in the segregated regions with those in the areas containing the purest dendrites, others with the mean composition of the steel.

The discussion is limited to a qualitative and necessarily very general comparison between the tendencies of the main elements to produce segregation, and an indication of some of the results obtained with Castaing's electronic probe.

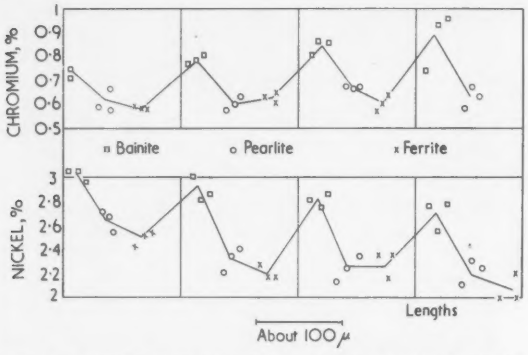

Fig. 8—Results of point analysis made with the Castaing microanalyser

Among the non-metals, carbon,[7, 8, 16, 25, 26] phosphorus,[7, 8, 16, 22, 26, 27] and sulphur[7, 8, 16, 26] can be considered to have a marked ability to produce segregation, but the diffusion of one can be greatly affected by the presence of the others, as well as by other metal constituents; this is particularly the case with carbon in the presence of phosphorus.

The existence of arsenic segregation can be demonstrated directly by autoradiography,[28] but the intensity of this segregation produced by different methods has not yet been clearly established[7, 8, 28]; in addition, there is no agreement on the possibility of homogenizing this type of segregation by a high-temperature heat-treatment.[22, 29]

Among the metals, manganese[16, 19, 20, 24, 27, 30] produces extensive segregation on the dendritic scale; copper[31] and tin,[31] molybdenum,[16] tungsten,[26] vanadium,[16] and probably cobalt also exhibit appreciable segregation. Nickel[7, 8, 16, 24] produces a fairly small dendritic segregation which, however, is clearly visible. Chromium[16, 34] and silicon[16] generally produce only slight dendritic segregation.

In general it is true to say that the impurities and the alloying elements in steel are as a whole more or less subject to dendritic segregation, to an extent which depends on the nature of the element and on its proportion in the steel, other conditions being equal. Carbon occupies a special place owing to its high diffusion coefficient which makes it easy to achieve its homogenization within the austenite by heating above the A_3 point, provided that it has not become fixed in the interdendritic spaces in the form of carbide by elements such as chromium, tungsten, vanadium, or molybdenum, and thus rendered rather difficult to redissolve. When the steel is cooled slowly through the Ar_3–Ar_1 range, this homogeneity of the carbon does not persist, the segregated elements locally modifying the transformation temperatures, or, more generally, the type of decomposition of the austenite.

We have tried to determine quantitatively the amount of dendritic segregation in a 100-ton ingot of a Ni–Cr–Mo steel of the following composition:

C, %	Si, %	S, %	P, %	Mn, %
0·24	0·30	0·024	0·015	0·53

Ni, %	Cr, %	Mo, %	Sn, %
2·5	0·65	0·27	0·30

A core of 17½ in. (435 mm) diameter was taken from the axis of a rough forging of diameter 7 ft 4 in. (1850 mm) produced from this ingot, and J. Philibert of IRSID investigated the segregation as shown by the very characteristic local heterogeneity of the microscopic structure, using Castaing's electronic probe. The results given here are those obtained for two areas only.

Area 1 (*bottom of ingot, surface of core*)

Microstructure: alternate light (ferrite) and dark (pearlite + ferrite) bands. The following results were obtained:

	Microstructure	Cr, %	Ni, %
First band	Ferrite + Pearlite	0·66–0·73 0·73–0·80	1·98–2·12
Second band	Ferrite	0·58–0·63	1·65–1·80
Third band	Ferrite + Pearlite	0·64–0·68 0·66–0·75	1·90–2·00 2·00–2·20
Fourth band	Ferrite	0·58–0·64	1·65–1·80

When these results are expressed as proportions, they indicate a segregation of 30% of both chromium and nickel; the difference in the nickel and chromium concentrations in the ferrites within the ferrite and the ferrite–pearlite bands is particularly interesting.

Similar results were obtained from measurements of the concentration of manganese and molybdenum; for example, the concentration of molybdenum ranges between 0·15% and 0·35%, with the following probable mean values:

Pearlite	0·3%
Ferrite	0·2%

Area 2 (*V segregates, 53% by weight from bottom of ingot*)

Microstructure consisting of three constituents arranged in two types of zone: a bainite zone and a zone containing ferrite and pearlite. The following results were obtained (Fig. 8):

Microstructure	Cr, %	Ni, %
Bainite	0·80–0·85	2·80
Pearlite + Ferrite	0·65–0·70 0·55–0·65	2·15–2·25

The more heavily quenched structure thus appears to be richer in special elements, with the relative segregation of nickel and chromium being about 35%.

The complete results, only a small part of which have been reproduced here, indicate that dendritic segregation is usually of the order of 30% for nickel, chromium, molybdenum, and manganese. In the large ingot used in this investigation they are thus of the same order as the large-scale segregations for the ingot as a whole, as determined by the chemical analysis of drill samples. Despite the fact that Castaing's probe analyses only a very small volume of material (a few cubic microns), there has been no evidence of any pronounced maxima in dendritic segregation which differ appreciably from the average values. The most heavily quenched constituent is always found to be richest in alloying elements.

Microsegregation always takes place on a very small scale, approximately 10 μ when a distinction has to be made between ferrite and pearlite in a pearlite–ferrite aggregate; this distance increases to 50 μ and

more for the case of an overall differentiation between bainite, pearlite, and ferrite.

THE ACTUAL MECHANISM OF FORMATION OF THE BANDED STRUCTURE

The mechanism of formation of banded structures caused by carbon segregation in the presence of phosphorus will be dealt with first, since this appears to play an essential part in the case of hypoeutectoid carbon and low-alloy steels.

The mechanism of the carbon migration may be described as follows. After solidification there are more carbon and phosphorus atoms in the interdendritic spaces than near the axes of the dendrites. It has been shown in pure iron–phosphorus alloys and in steels having increasing proportions of this element that the presence of phosphorus raises the temperature of the A_3 point at which austenite is transformed into ferrite (this effect is not shown by sulphur, for example). During a slow cooling, which allows the austenite to transform in the upper range, a steel with a relatively high phosphorus content will begin to become ferritic at a higher temperature than a steel containing less phosphorus. The same considerations apply to two separate areas of the same steel which differ in their phosphorus content as a result of dendritic segregation; ferrite will begin to form where phosphorus content is higher, while carbon becomes concentrated in those regions which are still austenitic.

The mechanism described above raises two important questions:

(*a*) *Why does the carbon diffuse away from the areas rich in phosphorus, and not the phosphorus?* The reason is that the activation energy of carbon for diffusion in austenite is smaller than that of phosphorus (43,000 and 32,000 cal/mol for phosphorus and carbon respectively), so that the carbon atoms diffuse more readily.

(*b*) *Why does the carbon migration result in an inversion of the heterogeneity and not in a uniform distribution of this element?* It has already been shown, especially by Rey,[32] that diffusion is controlled not by the concentration gradient but by an activity gradient. Numerous tests have indicated that the phosphorus reduces the solubility of carbon in alpha-iron, thereby increasing the thermodynamic activity of the former. Atoms thus diffuse from regions of high activity to regions of low activity, even if this leads to an increase in concentration (uphill diffusion).

Since the carbon migration is connected with the transformation from austenite to ferrite at A_3 in a heterogeneous metal, it is not surprising that the migration is not observed when the upper transformation disappears owing to sufficiently rapid cooling.

It is possible to extend these considerations still further and to apply them to practical cases. According to Grossmann,[33] phosphorus in solid solution produces a relatively large increase in the hardenability of the steel. But if there is both a rise in the A_3 transformation temperature and an increase in hardenability, the curves representing the start of the austenite transformation in steels of different phosphorus contents must intersect (Fig. 9).

Curve 1 shows, as a function of temperature, the time after which austenite begins to decompose in a very low-phosphorus steel; curve 2 applies similarly to a steel with a higher phosphorus content. These two curves intersect at x owing to the effect described above.

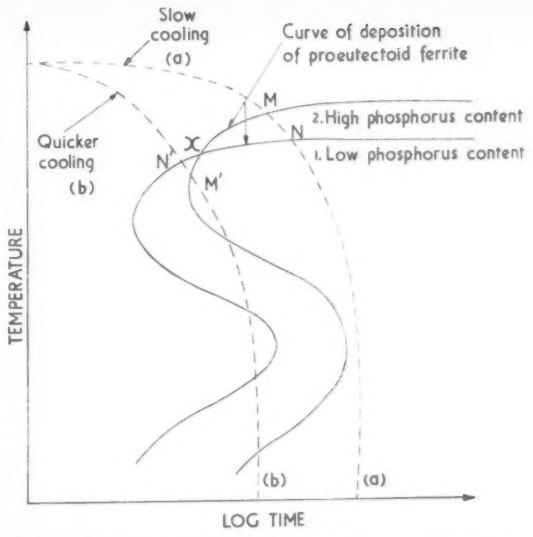

Fig. 9—Diagram showing conditions for eventual migration of carbon

It is obvious that the two curves could apply not only to two different steels but also to two different regions of the same steel: one corresponds to the dendrite axes (1) while the other is affected by segregation and is richer in phosphorus (2).

If cooling takes place slowly (curve *a*), as in the case of a reheat treatment, the cooling curve would intersect curve 2 at M, and curve 1 at N at a lower temperature. Ferrite is first deposited in regions with a higher phosphorus content, resulting in concentration of carbon in the remaining austenite.

With a more rapid rate of cooling, the corresponding cooling curve (*b*) intersects the transformation curves (1) and (2) at N' and M'. The austenite transformation now begins in regions with a low phosphorus content and tends to produce a concentration of the carbon in areas rich in phosphorus. The usual migration of the carbon away from phosphorus-rich areas will therefore not take place in this case.

These considerations lead to the conclusion that, from a certain cooling rate onwards, the regions with the highest phosphorus content can also be the last to begin the allotropic transformation, thus resulting in a concentration of the carbon atoms in regions richest in impurities. Provided that the composition of the steel is suitable, it should therefore be possible by means of a heat-treatment, followed by cooling at a suitable rate, to induce the carbon atoms to return to the original interdendritic spaces from which they have diffused during the preceding sufficiently slow reheat treatment.

These various cases were in fact found to be present during the comparative investigation into the banded structure of a Ni–Cr forging produced by acid, basic, or duplex methods,[7, 8] as described above.

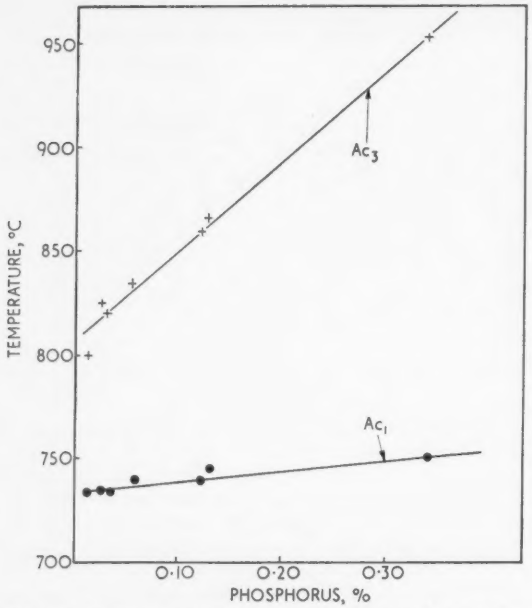

Fig. 10—Influence of phosphorus on the position of the Ac_1 and Ac_3 points

In order to provide an even more conclusive confirmation of the mechanism described above, it would be useful to establish experimentally the displacement of the transformation curves under the action of the phosphorus. This was in fact done in the Research Laboratories of the Le Creusot works in the course of a general study[35] of the banded structure of a steel of composition C, 0·28%; Ni, 1·56%; Cr, 0·52%; Al, 0·008%, to which phosphorus had been added to the extent of 0·025% (I), 0·056% (II), or 0·128% (III). The effect of the phosphorus on the temperature of the Ac_1 and Ac_3 points is shown in Fig. 10, which clearly indicates that the A_3 point is raised under the influence of the phosphorus.

Transformations under non-isothermal conditions were studied mainly during steady cooling and by comparing the steels containing 0·025% (I) and 0·128% (III) of phosphorus; this could be done very accurately by means of a Chevenard differential dilatometer, using instead of the standard bar a reference bar of steel I, and placing the sample of the steel to be investigated (steel III) in the usual position.[34] This arrangement gave very interesting results which have already been discussed elsewhere; it will be sufficient to note here that on cooling, the transformation in the test sample starts before that in the reference sample if the record curves upwards, and later in the test bar if the graph curves downwards (Fig. 11). Similarly, the transformation is completed in the test bar earlier or later than in the reference bar if the curve returns to the horizontal from above or from below.

Conventional dilatometer tests (samples I and III compared with a reference bar), as well as the more sensitive tests described above (sample I compared with sample III), have given the results shown in Fig. 11, which indicate the effect of phosphorus on the gamma → alpha transformation during cooling.

Table I
MEAN COOLING RATES, °C

	a	b	c	d	e	f	g	h	i	j
per hour	50	100	150	300	450					
per minute					7·5	13	50	90	225	
per second									3·75	12

The mean cooling rates between 850° C and the completion of the transformation are shown in Table I.

It is evident that the phosphorus raises the temperatures of both the beginning and the end of the transformation for the lower cooling rates *a–g*, whilst it lowers them for the very high cooling rate *j*. For an intermediate speed of cooling (*i*), the transformation starts first and finishes last with a steel having a high phosphorus content.

These experimental results thus confirm the basic principle of Fig. 9, namely the existence of a point of intersection *x* between the upper branches of the transformation curves of two steels with different phosphorus contents, and also possible mechanisms for separation of the proeutectoid ferrite described above.

A useful addition to this information is provided by a microscopic study of isothermally transformed steels, which indicates that the phosphorus does not appear to have any effect on the position of the M_s point (255° C), while an increase in the proportion of this element displaces the bainitic region towards the

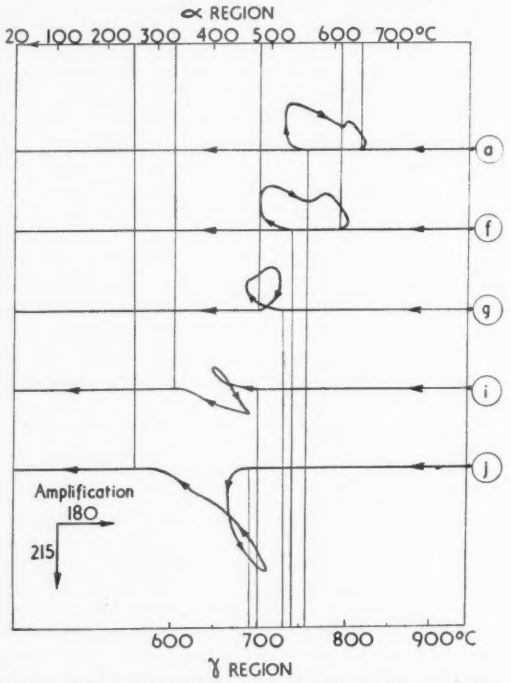

Fig. 11—Differential dilatometry of phosphorus steels. Reference specimen (I), P = 0·025%; comparison specimen (III), P = 0·128%

right and downwards. The beginning of the bainitic transformation is delayed slightly, and the actual transformation is slowed down considerably in the case of high phosphorus contents, mainly near 500° C.

These results lead to a better explanation of the various structures found in the bands. In the case of the two steels discussed above (I and III), the intersection x of the curves for the start of the austenite → ferrite transformation corresponds to a cooling rate of about 4° C per second. This eventually leads after a very limited ferrite separation to a structure which is almost entirely bainitic and very unsuitable for the occurrence of banded structures.

In other words, the increase in hardenability due to phosphorus in the steels investigated is too small and therefore fails to produce the conditions where the ferrite bands coincide with the original dendrite axes, the opposite of the usual result.

Effect of Dendritic Segregation of Elements other than Phosphorus

The conclusions reached above suggest that the various elements capable of producing dendritic segregation in steel can either reinforce or counteract the effect caused by phosphorus, as has already been proposed by various authors.[12, 13, 36]

According to Chipman,[37] the segregation tendency of an element L is proportional to the "segregation coefficient" $1 - k$, where

$$k = (X\%)_s/(X\%)_l$$

$(X\%)_s$ and $(X\%)_l$ being the proportions of the element in the solid and in the liquid, according to the Fe–L equilibrium diagram.

If Raoult's law applies both to the liquid and to the solid, and if the element is present in a low concentration, Chipman shows that

$$1 - k = \frac{M \times \Delta T}{1000\, X\%}$$

where M = atomic weight of element dissolved in iron
$X\%$ = proportion of element dissolved in iron
ΔT = lowering of melting point due to an $X\%$ concentration of element.

Values for $1 - k$ calculated by Chipman are shown below, together with those for arsenic and tin[35]:

C	P	S	O_2	As	Sn
0·87	0·87	0·98	0·98	~1	~0·7
Si	Mn	Cu	Ni	Cr	Mo
0·34	0·16	0·44	0·20	0·05	0·20

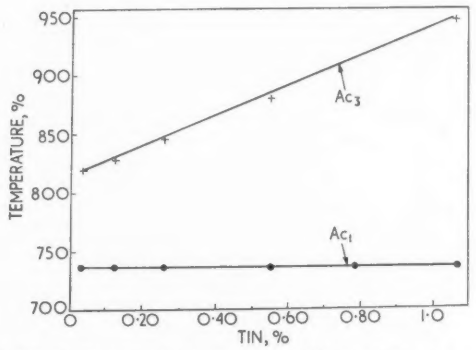

Fig. 12—Influence of tin on the position of the Ac_1 and Ac_3 points

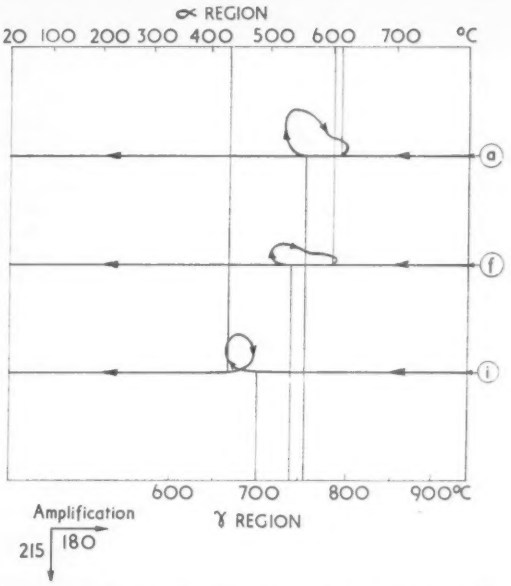

Fig. 13—Differential dilatometry of tin-bearing steels. Reference specimen (I), Sn = 0·035%; comparison specimen (III), Sn = 0·261%

The impurities in steel (phosphorus, arsenic, and tin) all suffer a very pronounced segregation and also have a similar effect on the temperature of the A_3 point, which they increase appreciably. In view of their slow diffusion in iron, it might be thought that they play a similar and essential part in the formation of the banded structure, particularly in the case of low-alloy or carbon steels.

In the course of an investigation at present in progress at the laboratories of the Le Creusot works, we have studied the effect of various tin contents (0·035%, 0·125%, and 0·260%) on a Ni–Cr steel of the same type as that used for the phosphorus investigation. The results obtained so far confirm that tin has a similar effect to phosphorus but to a lesser degree. It produces an increase in the temperature of the A_3 point (Fig. 12), but smaller than that found in the previous investigations; as a result the transformations from the austenitic state during cooling start earlier in a steel with a higher tin content (Fig. 13).

As early as 1918, Le Chatelier and Bogitch[38] suggested that the oxygen dissolved in steel plays a part in the formation of the banded structure by favourably influencing the separation of the ferrite; this suggestion was later taken up by Thompson and Willows.[10] It might appear that oxygen should have a very considerable effect on the ability to form segregations, since Chipman's coefficient $1 - k$ equals 0·98; it has, however, no effect on the A_3 transformation temperature and only a very limited solubility in iron. It is therefore quite probable that this element has only a very small effect, largely through the action of oxide inclusions.

The alloying elements nickel, chromium, and molybdenum, and also silicon, manganese, and

copper, suffer much less segregation than phosphorus, arsenic, and tin. Nickel and manganese reduce the temperature of the A_3 point and thus partly counteract the effect of the impurities discussed before. This may provide an explanation for the fact that, in certain special steels (which are also generally much freer from impurities), the annealed structures sometimes differ from those usually found in carbon steel.

Since metallic elements can readily be added in relatively large amounts, their segregation, although small in proportion, may nevertheless be appreciable in absolute terms. Various authors have attempted to explain the production of certain banded structures by manganese[39, 40] which lowers the temperature of the A_3 point and thus delays the formation of proeutectoid ferrite during cooling; the ferrite forms preferably in bands having a low manganese content. A structure of this type has been reported by Schwartzbart[13] in a steel containing 0·21% C and 1·47% Mn; slow cooling gave rise to the formation of pearlite bands in regions rich in manganese, and ferrite bands in areas with a low manganese content.

Inclusions in steel which have been arranged into laminæ through plastic deformation can, at least theoretically, play a part in the formation of the banded structure, either by favouring the physicochemical reactions at the interfaces, or by locally modifying the chemical composition of the steel and thus altering its transformation points if they are slightly soluble in iron at a high temperature. Any theories which ascribe to the inclusions an essential role in the formation of the banded structure[41] are, however, certainly incorrect; for a given steel and for different cooling rates, inclusions of a particular type may be found sometimes in the ferrite and sometimes in the pearlite. Other reasons for rejecting such theories have already been suggested,[10, 13] among them the effect of a homogenization treatment at a high temperature which partly destroys the banded structure without affecting the inclusions. We can therefore conclude that the inclusions have very little or no influence on the formation of banded structures.

CONCLUSIONS

The considerations outlined above lead to the conclusion that the banded structure in steel is produced by chemical heterogeneity due to dendritic or small-scale segregation during the cooling of the ingot. Each element which in solid solution produces a displacement of the temperature of the upper transformation point A_3 of the steel, accompanied in some cases by a horizontal displacement of the transformation curve (Fig. 9), can give rise to a banded structure in accordance with the conditions and the mechanism which we have attempted to describe.

Phosphorus plays an important part in the case of carbon and low-alloy steels and with bands of proeutectoid ferrite, while other impurities, especially arsenic and tin, can augment this effect. In general—and this applies especially in the case of alloy steels—most of the elements which can produce dendritic segregation influence the mechanism, either by reinforcing or by counteracting the effect of the impurities phosphorus, arsenic, and tin.

The action of carbon is of a special type. It diffuses so rapidly that its content can be assumed to be practically uniform throughout the metal in the austenitic state. During cooling through the transformation points, the effect of the segregated elements is either to reconstitute the original segregation of the carbon at the instant of solidification, or, on the contrary, to reverse this segregation. During certain transformations the increase in the carbon content of austenite which has not yet been transformed may be so marked that the final heterogeneity shown by the micrograph can be much more pronounced than that caused by the segregation proper.

A more detailed understanding of the mechanism of the formation of the banded structure can be very useful in increasing our knowledge of dendritic segregation and its results, in particular since some of the phenomena investigated are capable of very interesting generalization. For example, the ghost lines of large forging ingots often exhibit a migration of the carbon under the influence of segregated elements, in particular phosphorus, on the scale of the ghosts, which is similar to that shown on the dendritic scale by the banded structure.

Dr. Hatfield always showed a particular interest in the heterogeneity of the transformations of steel, and he made an important contribution to our knowledge in this field. I have therefore been very happy to present this Lecture in honour of his memory.

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Historical Note No. 2 (Second Series)—

An Early Invention for Protecting Iron against Corrosion

By H. R. Schubert

THE PROTECTIVE COATING of iron or steel with copper to prevent the access of corrosive agents has been the subject of investigation since the early part of the nineteenth century. It was achieved by immersing the metal at red heat momentarily in a hot bath of molten copper which formed a thin alloy layer over the surface.* In England this discovery was first made by a prince of the Royal House, Prince Rupert, who through his mother was a grandson of James I.

Prince Rupert was born at Prague in 1619, the son of the Elector Palatine, who was for a short time King of Bohemia. During the English Civil War the Prince served as a cavalry general and commanded a Royal army, but he was finally defeated by Cromwell's superior forces. From 1668 onwards he lived at Windsor Castle, where he had been appointed Governor; there the ingenious Prince devoted much attention to scientific studies and experiments conducted in a forge and laboratory which resulted in several important discoveries relating to the improvement of war materials.†

Prince Rupert is credited with the invention of malleable cast iron ("*softening cast or melted iron so that it may be filed and wrought as forged iron*"). This was patented in 1671, together with another of the Prince's inventions, the coating of iron with copper ("*tinctureing of copper upon iron*"). Four years later the patent was assigned to John Brown with authority to use Rupert's inventions for the making of iron ordnance, cast anchors, printing rolls, shells, and round shot.‡

John Brown was a member of a family of gunfounders who played an important part in the English iron and steel industry in the seventeenth century and operated several furnaces and forges, mainly on the border of Kent and Sussex.§ There is no doubt that he put Prince Rupert's inventions into practice.

Not long after Rupert's death in 1682 his inventions fell into oblivion, probably for want of skilled workers and because Brown was bound not to disclose the methods to anyone. Moreover, protective coating alone did not give more than a limited guarantee against corrosion, since coating effected by dipping alone could effectively resist corrosion only so long as the coating remained completely intact.

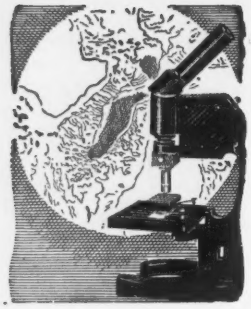
* R. M. Burns and W. W. Bradley, "Protective Coatings for Metals," pp. 235, 242 *et seq.*: 1955, New York, Rheinhold Publishing Corporation.

Prince Rupert, Count Palatine, 1619–1682
(Attributed to John Hoskins) [*National Portrait Gallery*]

† "Dictionary of National Biography," vol. XVII, pp. 405 *et seq.*, London, 1909; H. R. Schubert, "History of the British Iron and Steel Industry," pp. 270–271, 325–326: 1957, London, Routledge and Kegan Paul Ltd.

‡ County Record Office, Lewes, Sussex, Archives of the Hutton family.

§ E. Straker, "Wealden Iron," pp. 162 *et seq.*, 280, 281, 343, 1931, London; C. Ffoulkes, "The Gunfounders of England," p. 76: 1937, Cambridge.



PAPERS AND REPORTS ON RESEARCH AND PRACTICE

Metallurgy · Plant Operation · Research

Low-carbon Bainitic Steels

By K. J. Irvine, B.Sc., Ph.D., and F. B. Pickering, A.Met., A.I.M.

SYNOPSIS

A comprehensive investigation has been carried out on low-carbon bainitic steels as a result of which it is possible to produce weldable steels in the tensile strength range 40–80 tons/in². The main feature of these steels is that the normal polygonal ferrite formation is retarded so that transformation to bainitic ferrite can occur over a wide range of cooling rates. This is achieved most effectively with a ½% Mo–B base composition. The tensile strength of the steel depends to a large extent on the temperature of transformation, which is controlled by an additional alloy element. With additions of alloying elements up to about 3% it is possible to reduce the transformation temperature from 650° to 450° C and at the same time to increase the tensile strength from 40 to 80 tons/in².

Continuous-cooling diagrams have been determined and these indicate that the steels possess a very high bainitic hardenability so that consistent properties can be obtained in large section sizes up to about 24 in.

The tempering resistance is high and the steels only soften when ferrite grain growth occurs. Vanadium is particularly effective in increasing tempering resistance, and with 0·3% V present, good tempering resistance is obtained up to 750° C.

The combination of properties and transformation characteristics indicates that these steels should have a wide application as a high-strength weldable steel. 1378

Introduction

ALTHOUGH ATTENTION has been paid to intermediate transformation products in ferrous alloys since the early work of Davenport and Bain, essentially bainitic steels have found little commercial application. This has been partly due to the difficulties in obtaining such transformation over a reasonable range of section size, but also to a lack of knowledge of the physical metallurgy of such transformations.

One of the reasons why difficulty has been experienced in obtaining these structures is that bainitic structures are rarely obtained at air-cooling rates, except with very restricted section sizes. This is easily understood, because the transformation diagram of a plain carbon steel has the form shown in Fig. 1. In this type of diagram, the bainitic transformation region is shielded by the ordinary ferrite and pearlite transformation, and therefore over a range of cooling rates either ferrite and pearlite or martensite is formed.

When alloying elements are added, the ferrite and pearlite transformations are retarded but at the same time the bainitic region is depressed until it is quite close to the martensite transformation. This type of diagram is shown in Fig. 2 and it will be appreciated that, at air-cooling rates, it is extremely difficult to get essentially bainitic structures. The tendency is for mixed bainite–martensite structures to be formed too readily.

Another difficulty is the lack of information about intermediate transformations. It is not very easy to interpret isothermal transformation diagrams to

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Dr. Irvine is Head of the Metallurgy Section and Mr. Pickering is Senior Metallographer of the Research and Development Department, The United Steel Companies Ltd.

Table I

TYPICAL PROPERTIES OF STEELS ILLUSTRATED IN FIGS. 5*a–l*, NORMALIZED CONDITION

Part of Fig.	Steel No.	Strength, tons/in²					Elongation, %	Reduction in Area, %
		T.S.	0·2 P.S.	0·1 P.S.	0·05 P.S.	0·02 P.S.		
a	25	28·5	16·2 (YP)*	...	...	...	40·0	74·0
b	35	40·9	34·8	...	...	...	28·5	68·0
c	207	46·5	35·4	32·4	30·7	29·0	26·7	67·8
d	203	49·1	36·3	32·2	28·4	23·8	26·3	68·1
e	114	54·6	36·4	31·7	27·2	22·4	24·6	70·9
f	434	57·1	39·5	34·0	29·0	23·4	21·0	61·0
g	204	64·2	45·9	39·2	32·8	25·4	25·2	58·5
h	205	69·1	47·0	40·5	33·7	26·3	23·4	59·3
i	120	72·4	47·6	40·0	33·1	25·8	20·3	58·2
j	251	75·6	50·5	42·3	34·9	26·8	19·2	59·0
k	464	79·6	48·0	40·5	33·7	26·7	20·0	52·1
l	332	88·5	57·6	47·8	39·8	31·3	18·0	56·1

*Yield Point.

assess the relative proportion of transformation products which would be formed with slow continuous cooling, and also there are not many continuous-cooling diagrams which can be used. Those published by Steven and Mayer[1] relate to oil-quenching rates and do not extend to cooling rates which are slower than the air-cooling rate of a $\frac{3}{4}$-in. dia. bar. Liedholm[2] has published a number of continuous-cooling diagrams and recently Blickwede and Hess[3] also published diagrams for several standard A.I.S.I. steels. From these, it is possible to assess the proportion of transformation products that would be formed at air-cooling rates in thick sections. With A.I.S.I. 4340, air-cooling a 3-in. dia. bar would produce only 10% bainite with the remaining structure martensitic. Lower-hardenability steels, air-cooled in similar sections, would produce larger amounts of bainite, but these steels would also contain some ferrite and pearlite. Most previously published continuous-cooling diagrams, however, refer to alloy steel compositions of 0·3–0·4% C. There has been no reference to the transformation characteristics of steels containing 0·08–0·15% carbon and little discussion of the types of structure formed and the mechanical properties produced.

When low-carbon steels (0·08–0·15% C) are considered, the most important feature of the transformation diagram is the formation of pro-eutectoid ferrite which, except at very rapid cooling rates, is the major transformation product. This applies even when alloy additions are made, and a typical isothermal transformation diagram for this type of steel is shown as steel *A* in Fig. 3. This is a low-carbon, $\frac{1}{2}$% Mo steel, but a similar diagram would be obtained for any low-carbon, low-alloy composition. It will be noticed that the pro-eutectoid ferrite transformation still largely shields the bainitic transformation region. Consequently, except over a very limited intermediate range of cooling rates, a bainitic structure cannot be formed. Faster cooling rates will produce a martensitic structure whilst slower cooling rates will produce polygonal ferrite.

One recent investigation has been published, however, on the physical metallurgy of low-carbon, Mo–B steels.[4] When this composition is used a very marked change is made in the transformation characteristics. The normal formation of upper-temperature ferrite is hindered and steel *B* in Fig. 3 shows how this transformation is retarded. The effect of slowing down the formation of polygonal ferrite is to make possible the formation of a more acicular ferrite by a lower-temperature transformation. The transformation curve under continuous-cooling conditions is also shown in Fig. 3, and from this it will be appreciated that the transformation to acicular or bainitic ferrite will be obtained over a wide range of

Table II

EFFECT OF TEMPERING ON MECHANICAL PROPERTIES

Condition	Normalized	Normalized and Tempered 400° C	Normalized	Normalized and Tempered 400° C	Normalized	Normalized and Tempered 400° C	Normalized	Normalized and Tempered 400° C
Tensile strength, tons/in²	45·5	44·9	57·8	57·0	66·5	65·3	75·6	74·1
Proof stress, tons/in² for extension %: 0·2	33·8	34·2	41·8	46·1	46·4	52·9	50·5	58·3
0·1	31·6	32·6	35·8	42·7	39·6	49·1	42·3	53·0
0·05	30·0	31·8	29·8	39·9	33·1	46·0	34·9	48·9
0·02	28·4	30·9	22·8	36·6	27·0	42·5	26·8	44·7
Elongation on $4\sqrt{A}$, %	31·5	26·9	22·0	25·6	20·7	28·4	19·2	20·0
Reduction in area, %	68·9	70·6	54·7	69·2	65·3	68·7	59·0	62·4
Ratio $\frac{0{\cdot}2\%\ \text{P.S.}}{\text{T.S.}}$	0·74	0·76	0·72	0·81	0·69	0·81	0·66	0·78

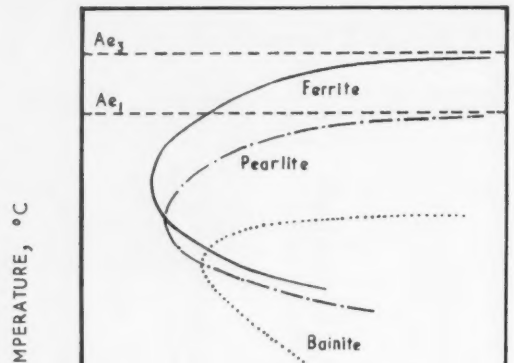


Fig. 1—Schematic representation of isothermal transformation diagram for a plain carbon steel, beginning of transformation only

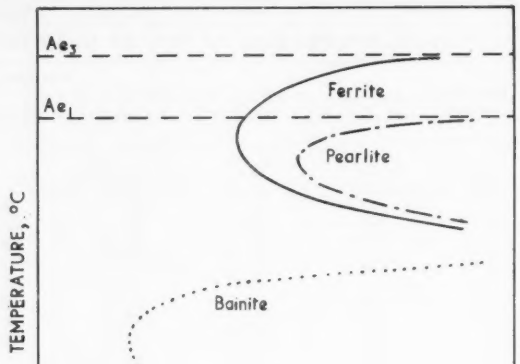


Fig. 2—Schematic representation of isothermal transformation diagram for a low-alloy steel, beginning of transformation only

cooling rates. The structure produced, which consists of fine-grained acicular ferrite together with randomly dispersed bainitic carbide aggregates, has considerably enhanced mechanical properties compared with a polygonal ferrite plus pearlite structure.

This work[4] on the Mo–B steels opened up a new field for investigation. In this type of steel the major transformation was occurring at about 630° C, even at slow cooling rates, and it seemed quite likely that further work of this type might enable an

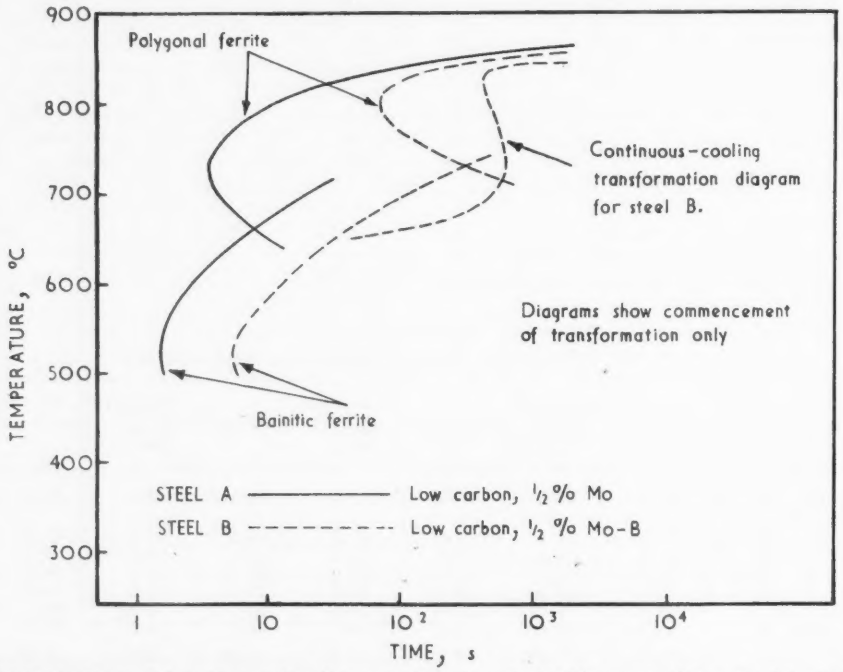


Fig. 3—Isothermal transformation diagrams for a low-C, ½% Mo steel with and without B

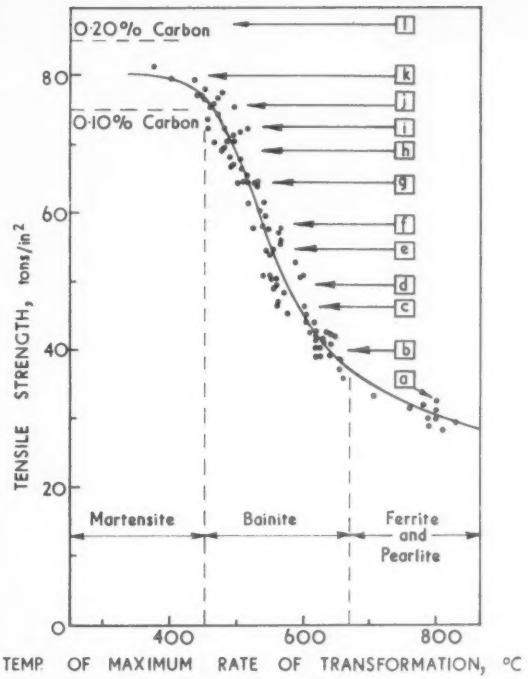
exploration to be made of the whole of the transformation range from 700° to 450° C, at which temperature the martensite transformation region is entered. This present paper describes the investigation which has been carried out and records the structures and properties which are obtained.

TRANSFORMATION CHARACTERISTICS OF LOW-CARBON BAINITIC STEELS

In the previous paper, the transformation characteristics of a low-carbon, ½% Mo–B steel were described. Briefly, the transformation to bainitic ferrite occurred over a range of about 80° C with the maximum rate of transformation at about 630° C. The mechanical properties of the structure produced by this transformation were about 40 tons/in² tensile strength and 30 tons/in² 0·2% proof stress. When a large number of steels to this composition had been tested, however, it was noticed that over the normal composition range it was possible to obtain tensile strengths varying from about 39 to 45 tons/in², and when a number of out-of-specification steels was included, this range was from 37 to 49 tons/in². All these steels had very similar metallographic structures, but it was noticed that there was a difference in the transformation temperature. If the temperature of the maximum rate of transformation was taken as the criterion, it was found that there was a relationship between tensile strength and transformation temperature. As the transformation temperature decreased, the tensile strength was increased. A transformation temperature of 640° C produced 38 tons/in², whereas when this was decreased to 580° C, the tensile strength increased to 48 tons/in².

Since it had already been suggested[4, 5] that the enhanced tensile strength of the Mo–B steel was due to the fine grain size of the low-temperature acicular ferrite it seemed logical that a still lower transformation temperature would produce a finer grain size with a consequently increased tensile strength. The decreased transformation temperatures appeared to be associated with increased contents of C, Mn, and Mo and this also seems a logical extension of the known effect of these elements on transformation temperatures. Consequently, a series of alloys was prepared with a base composition similar to that previously tested (i.e. 0·12% C, 0·5% Mo, 0·002% soluble B). To this base composition various amounts of alloying elements such as Ni, Cr, Mn, and Mo were added in amounts up to 3%. The cooling transformation characteristics and mechanical properties were determined on all these steels, and from the results a relationship was obtained between tensile strength and transformation temperature as shown in Fig. 4.

At this stage it is not wished to stress the individual compositions of these steels since it will be noticed that, irrespective of composition, the results fall in a reasonably narrow band, and it would appear that this relationship is a completely general one for transformations of this nature. At the lower end of the curve the transformation is to form polygonal ferrite and pearlite. At the top end of the curve, the transformation is just beginning to extend into the martensitic region. In between, i.e. at all transformation temperatures between 650° and 450° C, there is a



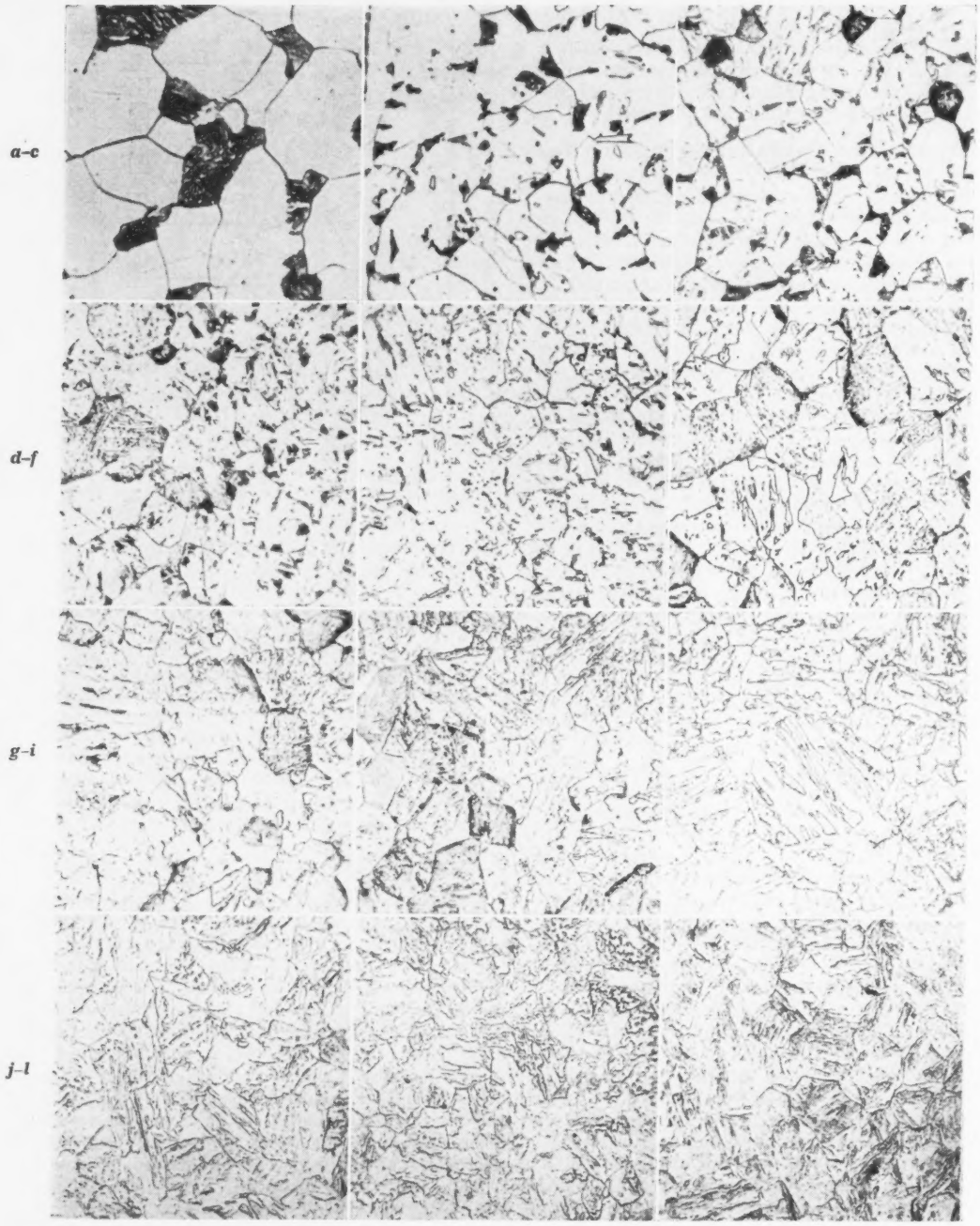
Letters refer to parts of Fig. 5.

Fig. 4—Effect of transformation temperature on mechanical properties (cooling rate for these results is equivalent to air-cooling ¾-in. dia. bar)

bainitic transformation. Although it will be discussed more fully later, it must be emphasized here that this bainitic transformation is obtained at a wide range of cooling rates equivalent to the air-cooling rates of a considerable range of section sizes.

METALLOGRAPHY OF LOW-CARBON BAINITIC STEELS

On the general transformation curve shown in Fig. 4 the location is given of the steels whose microstructures are shown in Figs. 5*a–l*. Figure 5*a* is that of a boron-free, ½% Mo steel and this structure is mainly polygonal ferrite with a small amount of pearlite. In the boron steels, at the lower end of the transformation range, the structure is typical of that previously described for the ½% Mo–B steel. The ferrite consists of small platelets which occur in groups of varying orientation, and bainitic carbides occur both between the plates and also adjacent to the prior austenite grain boundaries. As the transformation temperature decreases, the ferrite becomes more acicular and the carbide areas more dispersed. This effect continues until a transformation temperature of about 450° C is reached. Below this, the structure has the martensitic appearance shown in Fig. 5*l*. When it is remembered that the transformation of one of these steels occurs over a temperature range of about 80° C it will be realized that the final amounts of transformation of the steels shown in Figs. 5*j* and *k* are probably occurring below the calculated M_s point. No obvious appearance of martensite is noticed in these steels, but in all probability some of the very small particles



Figs. 5 *a–l*—Microstructures of steels located on general transformation curve in Fig. 4

shown in the microstructures are actually martensite areas. It is only when the carbon content exceeds 0·17% that any large martensitic areas are observed at this tensile level. This effect is probably a result of the known effect of higher carbon contents in retarding the bainite reaction.

Mechanical Properties

A very complete account of the mechanical pro-

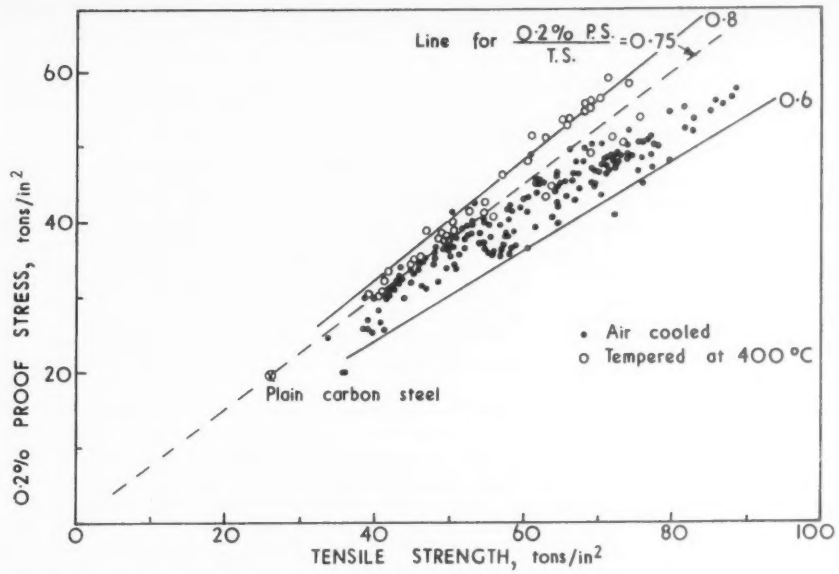


Fig. 6—Relationship between tensile strength and 0·2% proof stress, showing effect of tempering

perties of the ½% Mo–B steel* has been previously given.[6, 7] The most important point is that, compared to a non-boron, ½% Mo steel, the yield point is doubled. With typical properties of 40 tons/in² tensile strength and 30 tons/in² 0·2% proof stress, a PS/TS ratio of 0·75 is obtained. This is similar to the combination of properties obtainable in quenched and tempered steel and is in fact very desirable since the highest possible proof stress is obtained without approaching the dangerous state of having too little extension prior to failure.

The mechanical properties of the 12 steels whose microstructures are shown in Figs. 5*a–l* are given in Table I. It will be seen that the proof-stress values are increasing with the tensile strength, such that over the range of tensile strengths from 40–80 tons/in² the proof stress increases from 34–48 tons/in². At the lower end of the tensile-strength range, the PS/TS ratio of 0·75 is maintained, but when the test results for all the steels are plotted as in Fig. 6, it will be noticed that as the tensile strength increases, the ratio drops below 0·75, and, in fact, falls to 0·6. This is an important point since the proof-stress value may be of more importance in engineering design than tensile strength, and it might appear from these results that little benefit is obtained from going to the higher tensile strengths. In fact, however, this is an effect which results from the low transformation temperature of these steels and is rather similar to the effect noticed in quenched steels. A low transformation temperature produces residual stress which has the effect of lowering the proof-stress values, although the hardness and tensile-strength values are unaffected. Generally, it is possible to remove internal stresses by a low-temperature tempering treatment (of the order of 400° C), although this temperature is not sufficient to produce any tempering of the structure. A comparative series of mechanical properties after tempering at 400° C is given in Table II, and it will be noticed that the proof-stress values are now considerably increased. The 0·2% proof stress is increased nearly 10 tons/in² and the 0·02% proof stress is increased 20 tons/in² at the top of the property range. These tempering results are included in Fig. 6, and it will be noticed that all these results

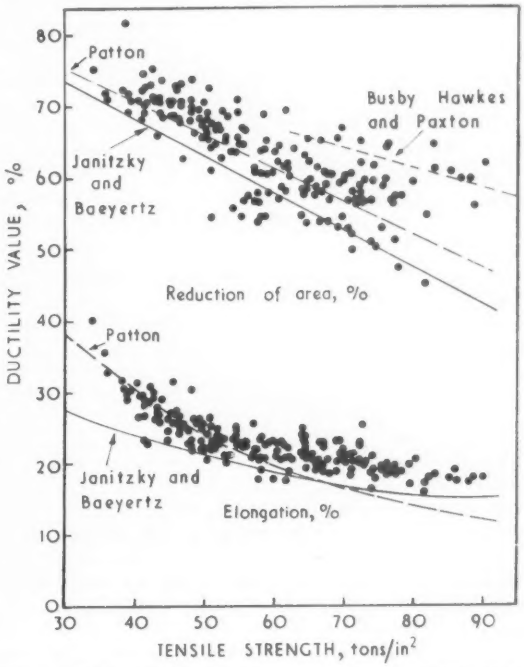


Fig. 7—Relationship between tensile strength and ductility values for bainitic steels

* This steel is covered by British Patent No. 611,598 and is registered as 'Fortiweld.' A patent of addition has been applied for to cover the steels in the higher tensile ranges dealt with in this paper.

fall in a band between 0·7 and 0·8 proof-stress ratio. With the one qualification, therefore, that at the higher tensile-strength values a tempering operation is necessary to produce the best combination of properties, the bainitic steels exhibit a satisfactory combination of properties over the whole of the tensile-strength range.

It is interesting to see how the properties of bainitic steels compare with those for quenched and tempered steels. When quenched and tempered steels have been considered, it has been shown that providing efficient quenching is obtained, the mechanical properties are little affected by the actual alloy content. There have been a number of summaries of such mechanical properties and Fig. 7 shows the lower tensile part of the summary curve drawn by Patton,[8] which included a large number of standard and NE compositions, and the curve drawn by Janitsky and Baeyertz.[9] The general feature of these curves is that, at least up to 80–90 tons/in², the mechanical properties form a fairly narrow band, and hence it is possible to predict the combination of associated mechanical properties which will be obtained at a particular tensile level. Generally speaking, the steels represented in these two curves are 0·3–0·4% C steels which have been quenched and tempered above 550° C.

The individual points in Fig. 7 represent typical bainitic steels between 40 and 80 tons/in². In general, the ductility values lie above the Patton and Janitsky–Baeyertz curves, particularly at the upper end of the tensile range (i.e. 55–80 tons/in²). In fact, at the higher tensile strength levels, the ductility values are approaching those obtained by Busby, Hawkes, and Paxton[10] for low-carbon martensites. These results confirm that when a comparison is made between low-carbon and high-carbon steels at the same tensile level, the low-carbon steel always shows superior ductility values. This applies to bainitic steels as much as to martensitic steels.

All the mechanical properties which are described in this paper have been determined in the normalized condition (as opposed to as-rolled). It should be pointed out, however, that almost identical properties would be obtained in the as-rolled condition. It has been found that finishing rolling temperatures between 1100° and 900° C have no effect on the mechanical properties, but a slight reduction in tensile strength is obtained if a very high temperature such as 1200° C is used, as in forging.

A very comprehensive investigation is being carried out on the impact properties of these bainitic steels, but this work will be reported separately. Briefly, however, it appears that the two most important factors controlling impact are tensile strength and grain size. At a constant tensile strength, the impact transition temperature is decreased if the grain size is reduced, whilst at a constant grain size the transition temperature is lowered when the tensile strength is reduced. To some extent this is self-regulating since the transformation temperature controls both the tensile strength and the grain size. If no special action is taken the grain size automatically becomes finer as the tensile strength increases and it is found that over the whole tensile-strength range, the impact transition temperature remains at about room temperature. Providing the initial grain size is fine, it is possible to reduce the transition temperature by tempering the steel to produce a slightly lower tensile strength. In this way it is possible to obtain transition temperatures below 0° C.

EFFECT OF ANALYSIS

The general feature of these steels is that a particular type of transformation has been produced by using a a ½% Mo–B base composition. With this base composition established, the effect of analysis has been investigated, covering C, Mn, Ni, Cr, Mo, W, V, and Ti.

It is already well established that carbon content has an important effect on the martensitic transformation, and therefore it would be expected that a similar marked effect would be shown on the bainitic transformation. To investigate this effect, four standard analyses were chosen. These were the ½% Mo–B base with additions of 0, ½, 1, and 1½% Cr. At the normal carbon content usually tested (0·12–0·14%) these steels would be expected to give roughly 42, 48, 55, and 63 tons/in² tensile strength and consequently the steels tested would fully cover the bainitic range. Carbon contents of 0·04, 0·08, 0·12, 0·16, and 0·20% were used and the combined effect of carbon and chromium variation had the marked effect on transformation temperature which is shown in Fig. 8*a*. At a constant chromium content the effect of raising the carbon content from 0·04 to 0·20% has been to reduce the transformation temperature by about 125° C. If the relationship shown in Fig. 4 is remembered, it will be expected that the reduction in transformation temperature will produce a very marked increase in tensile strength. It will be seen in Fig. 8*b* that this effect varies with chromium content and the increase is 12 tons/in² with a chromium-free steel and 32 tons/in² with a 1½% Cr steel. Since this effect is produced by a 0·16% increase in carbon content, 0·01% C produces an increase of 0·75 tons/in² in a chromium-free steel and 2 tons/in² in a 1½% Cr steel. It has already been mentioned that when the carbon content exceeds 0·17%, definite areas of martensite became visible in the microstructure at the higher tensile strengths. On the other hand, when the carbon content is very low the maximum tensile strength obtainable is reduced and it will also be shown later that the bainitic hardenability is reduced. An optimum range of carbon contents is therefore suggested as 0·08–0·14%.

Using a standard carbon content of about 0·12%, the effect of a number of alloying elements up to about 3% has been investigated. These effects are summarized in Fig. 9. On each of the sections of this diagram, the average curve (from Fig. 4) representing the relationship between transformation temperature and tensile strength has been drawn, and then there is also included the individual points obtained from the steels having a variation in a single alloying element. Generally speaking, these single points fall very closely on the average curve, but it will be noticed that particularly in the case of Ni and Mn, the points tend to fall below the average curve. This effect is due to the fact that each alloying element

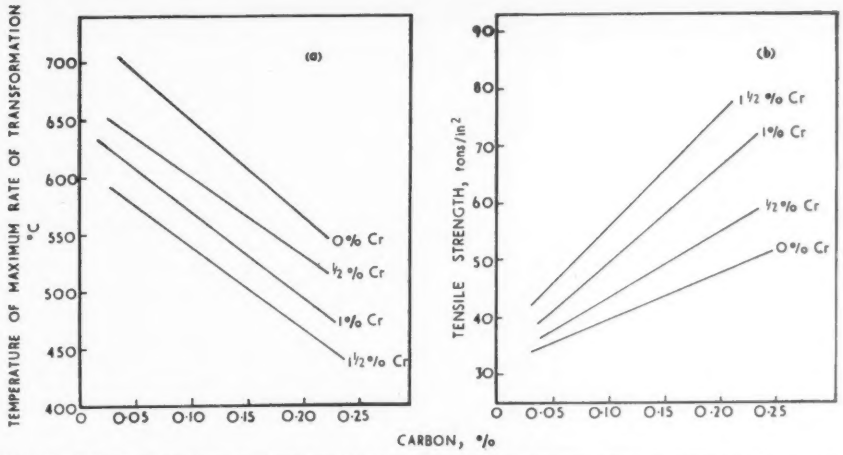


Fig. 8—Effect of carbon content on (*a*) transformation temperature at varying Cr levels, (*b*) tensile strength at varying Cr levels

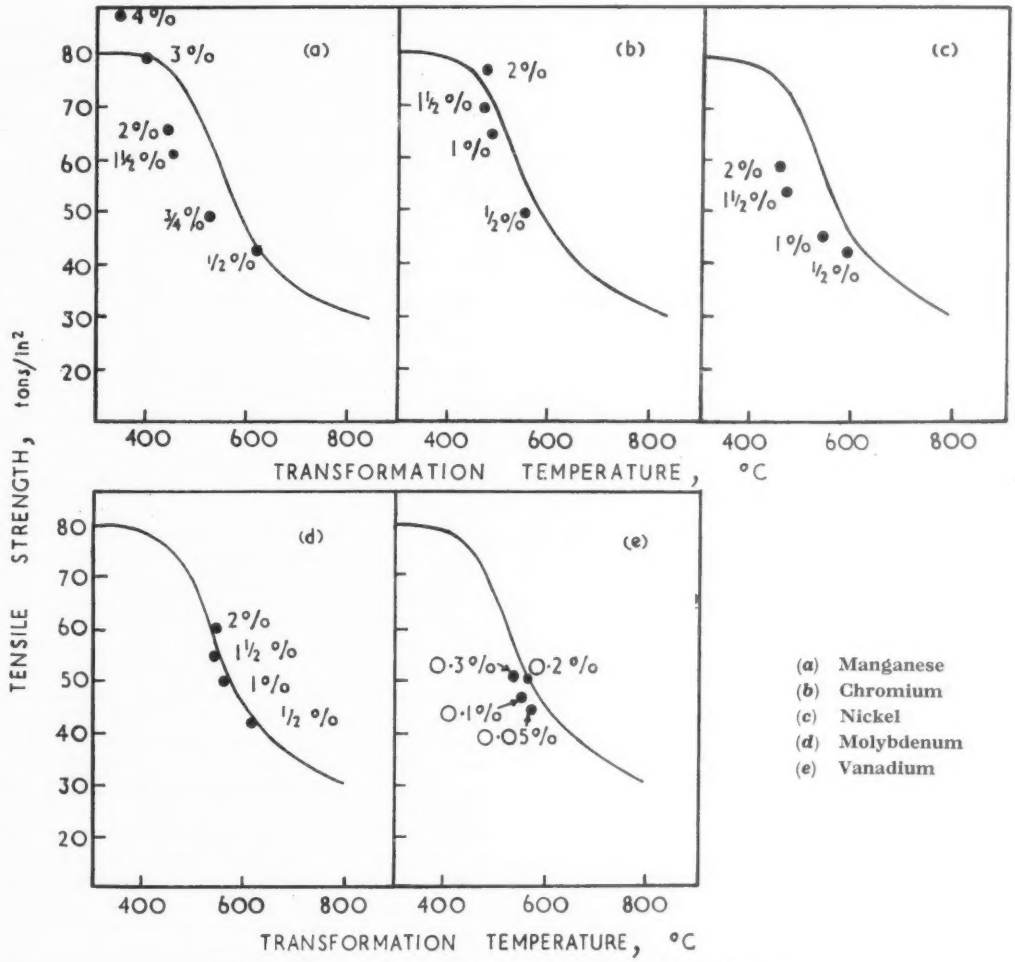
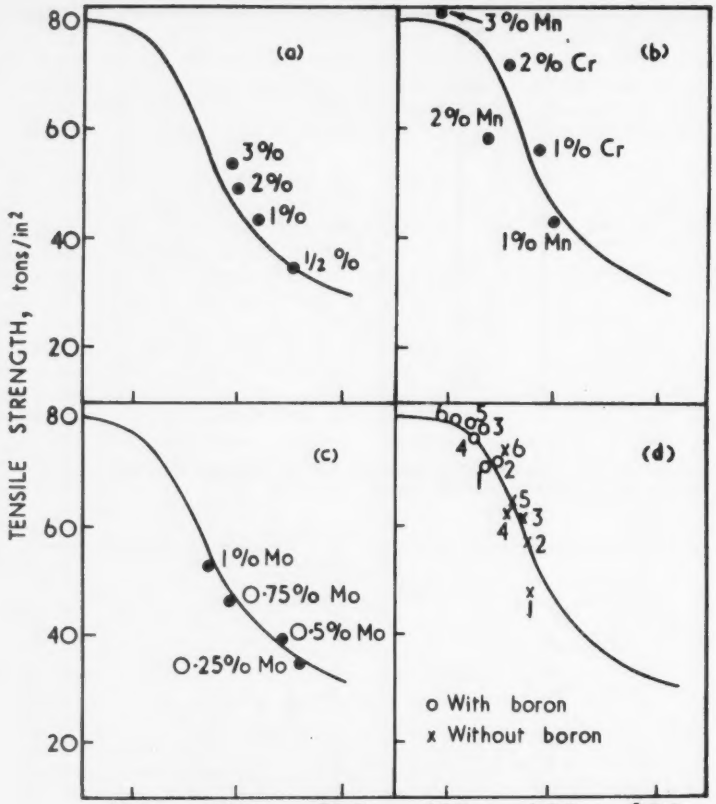


Fig. 9—Effect of alloy additions to a 0·12% C, ½% Mo–B steel



(*a*) W in B steels (*b*) Mn and Cr in 1% W–B steel
(*c*) Effect of Mo in 1% Cu–1% Ni steel (*d*) Effect of B

Fig. 10—Alternative base compositions for bainitic steels

probably has two separate effects. There is the effect which depresses the transformation temperature and there is a solid-solution hardening effect which will be related to the size of the alloy element atom. Consequently, it would appear that Mn and Ni have a small solid-solution hardening effect, whereas Mo and V have a much greater effect. In the case of V and Ti, an increasing addition raises the tensile strength until secondary effects such as combination with carbon or ferrite formation change the basic structure.

A summary of the effect of the individual alloying element can be given as follows:

Alloying element, 1%	Tensile strength, tons/in²
Mn	15
Cr	15
Ni	8
Mo	12
Ti	25
V	25
Cu	4

At an equivalent tensile strength, the steels have a similar microstructure, irrespective of chemical composition, and these structures are the same as those shown in Figs. 5*a–l*. They are all bainitic structures with the exception of the 4% Mn steel. This has a typically martensitic appearance and Fig. 9*a* shows that this steel occupies a position well above the general bainitic transformation curve.

CHANGE IN BASE COMPOSITION

In all the steels which have been described, a common base composition was used, viz. ½% Mo–B. The purpose of this base composition was to retard the high temperature ferrite formation and expose the bainitic transformation so that bainitic structures are obtained at air-cooling rates. In the background work to the ½% Mo–B steel, it was shown that it was possible to replace Mo by W, although a larger amount was needed (1% W to replace ½% Mo).

In order to investigate the effect of a change in base composition a series of steels was tested using a 1% W–B base. Figure 10*a* shows the experimental results obtained with various W additions. These lie close to the general curve obtained for ½% Mo–B steels. If the results are compared with those in Fig. 9*d* it will be seen that about twice as much W as Mo is needed to produce an equivalent effect. On this basis, the equivalent value for W is 1% W = 6 tons/in² in the bainitic transformation range.

A low-carbon, 1% W–B steel has a very similar transformation diagram to a ½% Mo–B steel and, once the transformation characteristics have been regulated to allow bainitic structures to form at air-cooling rates, further alloy additions produce the same lowered transformation temperature and increase in tensile strength, as in a ½% Mo–B steel. Figure 10*b* shows the effect of manganese and chromium additions to the 1% W–B base. The experimental points lie on the general curve and the same equivalents apply as to a ½% Mo–B base.

Other work has previously been published on molybdenum steels. An early patent[11] described a low-carbon, Mo–V–Cu steel which gave high tensile strengths in the air-cooled condition. These steels were undoubtedly bainitic steels, although this was not realized at the time, and a suggestion was made of a precipitation-hardening mechanism. Some interesting work has been carried out by Nehl and Rose.[12] In the course of their efforts to develop a high-tensile steel, they noted that a 1% Cu, 1% Ni, 0·35% Mo steel possessed good strength and excellent welding characteristics. They determined continuous-cooling transformation diagrams and pointed out that it was a bainitic transformation which was producing the good mechanical properties.

Table III
EFFECT OF BORON

Steel No.	Identiffcation in Fig. 10*d*	Alloy Addition to 0·12–0·14% C, 0·5% Mo Base		Tensile Strength	L.R.C., °C/min	Critical bar dia., air-cooled, in.
		Mn	Cr			
459	1	0·8	1·0 + B	70·3	30	5
460	2	0·8	1·5 + B	71·5	25	7½
461	3	0·8	2·0 + B	77·5	15	14
462	4	1·5	1·0 + B	77·1	15	14
463	5	1·5	1·5 + B	78·1	10	19
464	6	1·5	2·0 + B	79·6	< 6	> 24
		Boron-free				
465	1	0·8	1·0	46·8	180	¾
466	2	0·8	1·5	57·8	80	2
467	3	0·8	2·0	63·8	50	3
468	4	1·5	1·0	64·6	35	4
469	5	1·5	1·5	64·5	30	5
470	6	1·5	2·0	75·5	30	5

A more recent patent[13] describes a low-carbon, Mn–Ni–Cr–Mo–V–Cu steel, and a description of these steels[14] together with microstructures indicates that this composition also gives an essentially bainitic structure, at least in thinner sections, with tensile strengths up to 70 tons/in².

It seems clear that non-boron compositions can give bainitic structures and that a Cu–Mo base is one possibility. Since Ni is usually added to copper-bearing steels for improved hot-working properties a series of low-carbon, Cu–Ni–Mo steels was tested. These contained additions up to 1% of these three elements added to a 0·12% C base. The experimental results for transformation temperature and tensile strength are indicated in Fig. 10c. The structure is bainitic and the mechanical properties are reasonable, but further work has shown that the bainitic hardenability is extremely limited.

The other alloy compositions mentioned (Mn–Ni–Cr–Mo–V–Cu) indicate that boron-free steels can certainly produce bainitic steels having tensile strengths of 70 tons/in² (in thinner sections). A number of experiments has therefore been carried out with similar steel compositions with and without a boron addition. The base compositions and tensile strengths obtained are given in Table III. It will be seen that at the lowest alloy content, the omission of boron reduces the tensile strength by nearly 25 tons/in² whereas at the highest alloy content, there is only a drop of 4 tons/in². Figures are also given for the L.R.C. value, which is the rate of cooling at which high-temperature ferrite formation occurs, and these figures are also converted to the maximum bar diameter which would still have a bainitic transformation at the centre when air-cooled. These figures show that boron is contributing very much to bainitic hardenability, particularly at the lower strength levels. At the lower strength levels, the boron-free steels are too limited in section to be of any interest.

From these results it can be concluded that with a typical alloy addition of about 0·8% Mn and 1% Cr, an addition of 0·002% soluble boron is equivalent to an increase in tensile strength of about 25 tons/in². Alternatively, another way of regarding this effect is to say that a 0·8% Mn, 1% Cr, ½% Mo–B steel is equivalent to a 1·5% Mn, 2% Cr, ½% Mo, boron-free steel. Consequently, the addition of 0·002% soluble boron is equivalent to the addition of about 0·7% Mn and 1% Cr.

An additional effect of the boron addition is the progressive increase in bainitic hardenability with high alloy contents. Although the boron generally raises the tensile strength by means of a reduction in transformation temperature, the increase in bainitic hardenability indicates that this is being accomplished not so much by a depression of the transformation line as by a general shift to the right on the transformation diagram, i.e. to slower cooling rates. The critical bar diameter increases to over 24 in. and this effect of boron has very great industrial significance for large-section applications such as die blocks and rotors.

CALCULATION OF TENSILE STRENGTH

As the effect of each of the individual alloying elements in increasing the tensile strength in the bainitic range has been established, it is possible to produce an equation which enables the tensile strength

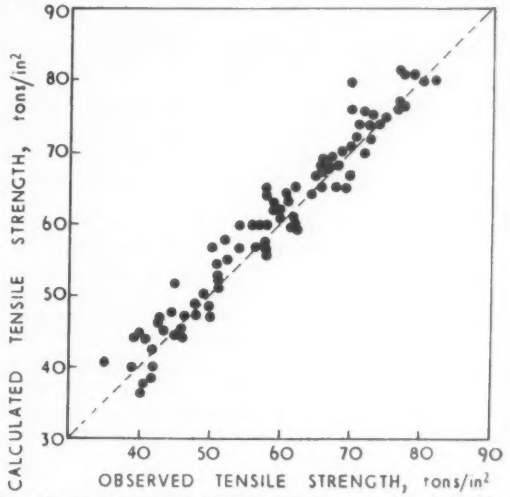
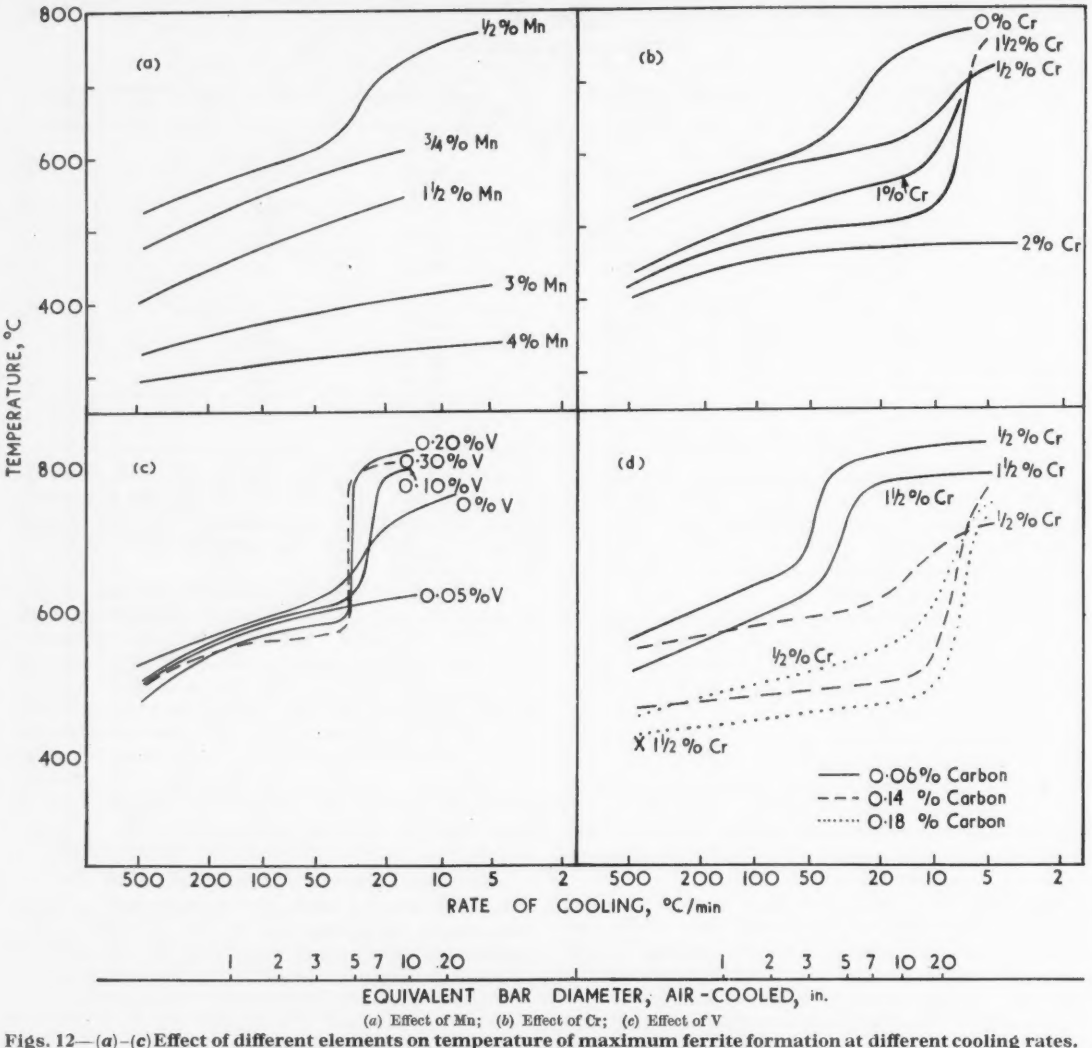


Fig. 11—Relationship between observed and calculated tensile strength



(*a*) Effect of Mn; (*b*) Effect of Cr; (*c*) Effect of V

Figs. 12—(*a*)–(*c*) Effect of different elements on temperature of maximum ferrite formation at different cooling rates. (*d*) Effect of C and Cr on continuous-cooling transformations

of a mixed alloy composition to be calculated. Carbon is the most difficult element to assess since its effect varies at different strength levels. At the bottom of the tensile range 0·01% C = 0·25 tons/in², whilst at the top 0·01% C = 2 tons/in².

In the middle of the tensile range and over the normal range of carbon contents (0·08–0·15%), however, 0·01% C = 1·25 tons/in². Therefore, a factor of 1% C = 125 tons/in² has been used for the equation. There is a large amount of evidence to show that a base composition of 0·12% C, 0·5% Mn, 0·5% Mo + B has a tensile strength of 42 tons/in². Using the equivalents which have been given previously for the individual alloying elements the tensile strength of a higher alloy bainitic steel can therefore be calculated from the formula:

$$\text{T.S.} = 42 + (\text{C}-0{\cdot}12)125 + (\text{Mn}-0{\cdot}5)15 + (\text{Cr})15 + (\text{Mo}-0{\cdot}5)12 + (\text{W})6 + (\text{Ni})8 + (\text{Cu})4 + (\text{V} + \text{Ti})25.$$

It is also possible to use a 1% W–B base in which case the equation is

$$\text{T.S.} = 42 + (\text{C}-0{\cdot}12)125 + (\text{Mn}-0{\cdot}5)15 + (\text{Cr})15 + (\text{Mo})12 + (\text{W}-1{\cdot}0)6 + (\text{Ni})8 + (\text{Cu})4 + (\text{V} + \text{Ti})25.$$

However, the more general case is for the ½% Mo–B base, and the first equation becomes

$$\text{T.S.} = 16 + 125\,\text{C} + 15\,(\text{Mn} + \text{Cr}) + 12\text{Mo} + 6\text{W} + 8\text{Ni} + 4\text{Cu} + 25\,(\text{V} + \text{Ti}).$$

Using this formula, the tensile strength has been calculated for nearly 100 steels on which tensile data has been determined. In Fig. 11, the calculated tensile strength is plotted against the actual tensile strength. It will be seen that there is close agreement between these two values. For the majority of the steels the values agree to ± 5 tons/in² and this is maintained over the complete tensile-strength range. The calculated values in general lie slightly above the line of perfect agreement by about 2 tons/in². This

Table IV — EFFECT OF COOLING RATE ON MECHANICAL PROPERTIES OF BAINITIC STEELS

Alloy Addition to 0·14% C, ½% Mo–B base	½% Cr							1½% Cr							2% Cr						
Cooling rate, ° C/min	450	180	100	32	22	12	6	450	180	100	32	22	12	6	450	180	90	32	22	12	6
Tensile strength, tons/in²	50·2	47·2	48·0	46·3	44·8	43·2	36·4	65·6	66·5	63·7	63·7	64·0	61·4	32·6	75·8	72·8	74·0	72·4	71·4	69·6	69·6
Proof stress, tons/in² for extension %																					
0·20	32·9	31·5	33·2	32·9	32·7	30·6	32·0	43·5	43·5	42·3	42·8	41·4	40·5	...	50·4	47·8	49·3	47·3	46·6	46·6	44·8
0·10	29·1	27·6	29·0	28·5	29·4	27·7	18·7	36·7	37·5	35·8	36·3	36·7	34·2	18·9	41·4	40·2	41·0	39·8	39·1	39·8	37·5
0·05	24·4	23·7	25·4	24·9	26·7	25·3	16·4	29·9	31·1	29·9	30·4	31·1	28·4	YP*	32·1	33·3	33·4	32·9	32·5	34·0	31·2
0·02	19·3	19·3	21·1	20·4	23·9	22·7	13·7	23·0	24·1	23·0	23·4	25·2	22·4	...	24·5	25·5	25·3	25·1	24·4	28·2	24·9
Elongation on $4\sqrt{A}$, %	25·1	26·8	24·5	26·8	28·7	29·2	34·8	21·4	23·8	22·0	20·7	19·7	21·4	41·4	19·6	19·4	21·1	22·4	21·4	23·3	21·3
Reduction in Area,	64·6	64·6	66·8	69·2	74·8	73·6	71·4	63·7	64·6	63·2	59·8	59·8	62·2	73·6	59·8	59·8	59·8	62·2	57·2	54·6	66·8

*Yield point

could be rectified by altering the constant in the equation, but it is felt that if more accuracy is required then closer attention should be paid to the carbon factor, which should probably vary with different carbon ranges. It should be pointed out that the steels included in Fig. 11 covered a range of 0·05–0·20%.

CONTINUOUS COOLING TRANSFORMATION CHARACTERISTICS

(*a*) Effect of Analysis on the Continuous Cooling Transformation

Series of steels have been investigated using cooling rates ranging from about 6–450° C/min. These steels contained up to 2% Cr, 4% Mn, and 0·3% V added to the ½% Mo–B base composition. The effect of variation in carbon content from 0·06–0·18% has also been studied. The temperature of maximum rate of transformation was used to draw the transformation diagrams and these are shown in Figs. 12*a–d*. There are three important effects of the alloying elements which are shown by these diagrams.

(i) The depression of transformation temperature, since this will govern the increase in tensile strength
(ii) The increase in bainitic hardenability, or the L.R.C. value for a particular composition, since this determines the maximum section size in which full bainitic properties will be obtained
(iii) The slopes of the bainite transformation line since this determines the variations in properties which will be obtained in different section sizes.

It will be seen from Figs. 12*a* and *b* that the general effect of both Mn and Cr is to depress the transformation temperature, to increase the bainitic hardenability, and to decrease the slope of the bainite transformation line. Consequently, an increase in both these elements will raise the tensile strength and produce more uniform properties over a wider range of section sizes. Chromium is slightly more effective than Mn, and 2% Cr added to the base composition produces a very flat transformation line with the major transformation to bainitic ferrite even at the slowest cooling rate used (6° C/min). Vanadium has a slight effect (Fig. 12*c*), and in fact with more than 0·1% V present, the bainitic hardenability is restricted, probably owing to the presence of δ-ferrite or a reduction of the effective carbon content due to the formation of vanadium carbide.

Figure 12*d* shows that carbon is very effective in lowering the transformation temperature and increasing the bainitic hardenability. This applies irrespective of the remaining alloy composition, and with 0·06% C the bainitic hardenability is low, even with 1½% Cr added. Consequently, it is suggested that the minimum carbon content of these bainitic steels should be 0·08%.

(*b*) Effect of Rate of Cooling on the Mechanical Properties

The effect of rate of cooling on the mechanical properties can be illustrated by the series of steels containing chromium additions. The mechanical properties for these steels are given in Table IV and it will be noticed that as the chromium content increases, the general level of tensile strength increases and the variation in mechanical properties decreases. When the chromium content reaches 2% a quite remarkable consistency in properties is obtained. At cooling rates equivalent to air-cooling of ¾-in. to 24-in. dia. sections, not only is the tensile strength high (about 75 tons/in²), but there is only a variation of 6 tons/in² over the complete range of cooling rates.

(*c*) Effect of Rate of Cooling on Microstructure

The effect of rate of cooling on transformation can be followed quite clearly from the microstructures. Slower rates of cooling raise the transformation temperature and hence produce less acicular microstructures. When the limiting rate of cooling is reached, polygonal ferrite can be easily detected. This effect is shown in Figs. 13*a–e* which shows the structures obtained in a 1% Cr bainitic steel at various rates of cooling. At 180° C/min the typical structure expected from a steel of this tensile strength (65·6 tons/in²) is obtained (Fig. 13*a*). When the cooling rate is reduced to 100° C/min, the unchanged tensile strength (66·5 tons/in²) is matched by a similar structure (Fig. 13*b*). This effect continues until a cooling rate of 12° C/min is reached (Figs. 13*c* and *d*). This is close to the limiting rate of cooling, and the slower rate of 6° C/min produces a marked drop in tensile properties (32·6 tons/in²) and polygonal ferrite can be seen in the microstructure (Fig. 13*e*).

TEMPERING CHARACTERISTICS

The tempering characteristics of bainitic steels are important because of the possible use of tempering in

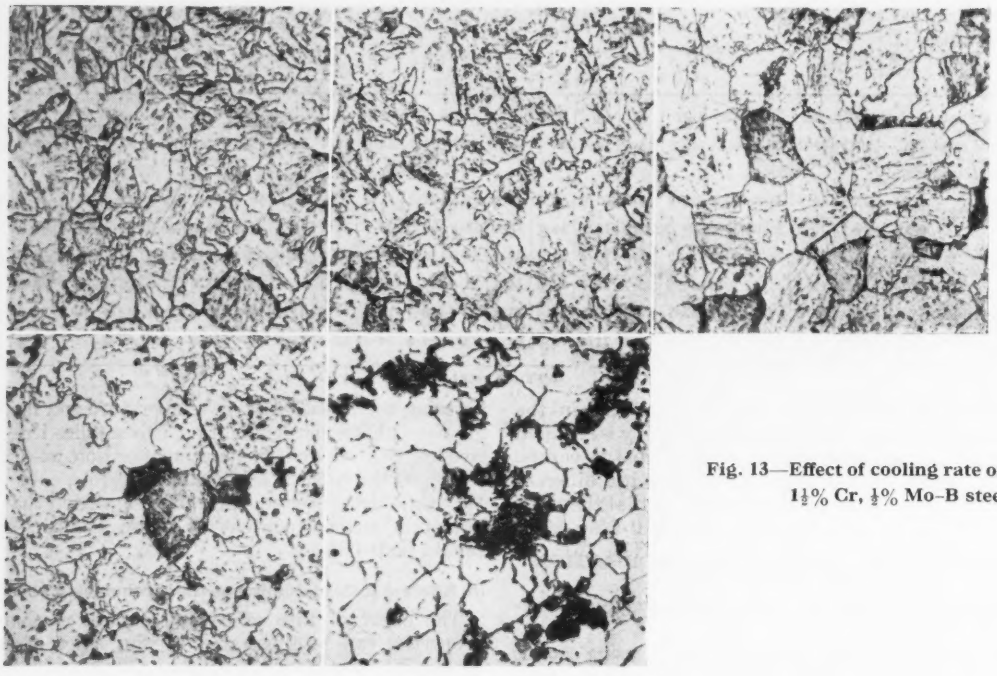


Fig. 13—Effect of cooling rate on 1½% Cr, ½% Mo–B steel

commercial heat treatment and also because these steels may have some high-temperature applications. The results of the tempering tests which have been carried out are also interesting since they provide further information about the bainitic structure itself. The tempering characteristics have been determined for a number of bainitic steels containing various alloy additions. The change in hardness has been expressed in terms of a combined time–temperature parameter, $T°\,K\ (20 + \log t$ hours). A typical tempering curve is shown in Fig. 14 which represents the 0·14% C, ½% Mo–B base composition, and the general form of the tempering curve was shown by all the steels. The hardness is maintained uniform until a tempering parameter of 18,000 is reached (equivalent to 1 h at 625° C or 100 h at 540° C). Above this there is a progressive softening with increasing tempering parameter.

The effect of Mn is shown in Fig. 15*a*. Manganese contents of 0·85% and 1·42% were added to the base composition (which contains 0·5% Mn). It will be seen that apart from the difference in initial hardness, Mn has no effect on the form of the tempering curve. Figure 15*b* shows that Cr has a similar lack of effect. Additionally in this series, a carbon content of 0·06% was used, and a comparison between Figs. 15*a* and *b* shows that carbon also has no effect on the form of the tempering curve. In contrast, 0·3% V introduces a marked secondary hardening peak. Figure 15*c* shows that this effect is not obtained with 0·1% V, so it is likely that a critical vanadium content or more probably a critical V/C ratio is necessary for secondary hardening to occur. There are two 0·3% V steels shown, only one of which contains boron. The similar form of these two tempering curves indicates that boron plays no part in the tempering

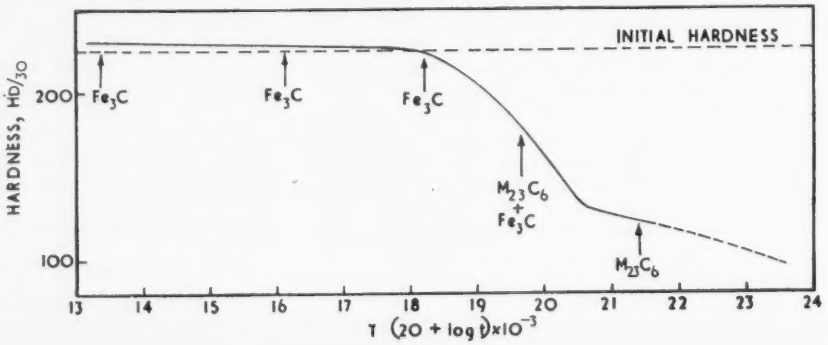
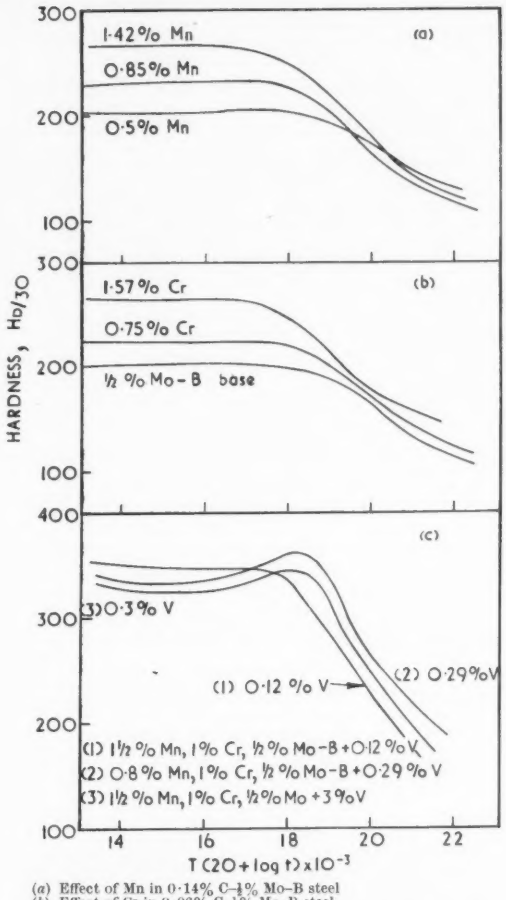


Fig. 14—Tempering characteristics of a bainitic steel

Analysis: 0·14% C, 0·85% Mn, 0·50% Mo, 0·002% soluble B



(*a*) Effect of Mn in 0·14% C–½% Mo–B steel
(*b*) Effect of Cr in 0·06% C–½% Mo–B steel
(*c*) Effect of V

Fig. 15—Effect of analysis on tempering characteristics of bainitic steels

reactions, a fact which has been noted previously. The secondary hardening effect of 0·3% V increases the tempering parameter for initial softening from 18,000 to 19,000 (equivalent to 1 h at 685° C or 100 h at 590° C).

These tempering curves have therefore established that the low-carbon bainitic steels are extremely resistant to tempering as shown by the extreme case of the 0·3% V steel (initially 70 tons/in²) which after 100 h at 700° C still has a hardness equivalent to a tensile strength of 40 tons/in².

(*a*) Microstructural Changes during Tempering

The structural changes can be related very easily to the tempering curves. Typical microstructures are shown in Figs. 16*a–f* for the 1½% Mn steel (initial tensile strength, 60 tons/in²), and in Figs. 17*a–f* for the 0·3% V steel (initial tensile strength, 70 tons/in²). In Figs. 16*a–f*, the change in microstructure can be compared with the tempering curve shown in Fig. 15*a*. The structure changes very little at low tempering parameters (i.e. up to 18,000). At higher tempering parameters, a structural change accompanies the fall in hardness. There is a coalescence of carbide particles and a grain growth or recrystallization to produce large polygonal grains of ferrite.

There is a noticeable change in the tempering characteristics of the vanadium steel shown in Figs. 17*a–f*. These structures should be compared with the tempering curve in Fig. 15*c*. The structure shows little change at low tempering parameters and in fact loses much of the definition which is given to it by carbide particles. These seem to re-enter solution during the secondary hardening stage (Fig. 17*c*). At higher tempering parameters (above 19,000) coalescence of precipitated carbide particles occurs at both the prior austenite and ferrite plate grain boundaries, but even with this change there is no massive grain growth to form polygonal ferrite grains. The ferrite retains much of its acicularity and hence, to some extent, its strength (Figs. 17*c* and *f*).

(*b*) Carbide Formation

X-ray identification of extracted residues has been used to study carbides present in specimens at various positions on the tempering curves.

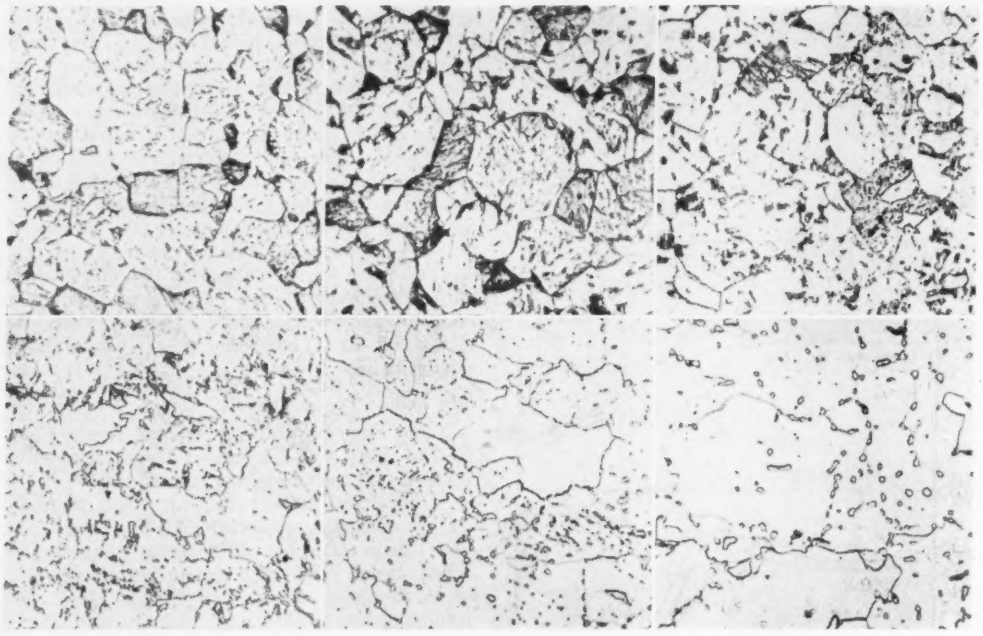
At low parameter values at which there is little change in hardness, the carbide is Fe_3C and tends to be platelike in form, but with further tempering, coalescence occurs. The basic carbide reaction accompanying softening is $Fe_3C \rightarrow M_{23}C_6$, the $M_{23}C_6$ growing rapidly as the Fe_3C re-enters solution and allows grain growth and softening to occur. The $M_{23}C_6$ carbide is based on $Fe_{21}Mo_2C_6$, the published lattice parameter of which is 10·51 kX, compared with the measured value in the low-manganese steel, of 10·516 kX. It is assumed that there would be no other alloying element than Mo (and maybe a little Mn replacing Fe) in the carbide, so the agreement in lattice parameters is quite good.

Chromium appears to enter the $M_{23}C_6$ carbide by simply replacing Fe to give $(FeCr)_{21}Mo_2C_6$ and this causes an expansion of the lattice. The element which is really controlling the carbide type developed, however, is Mo and no separate chromium carbide has been found up to 1½% Cr. In the high-manganese steels, Mn does appear to enter the $M_{23}C_6$ and causes a decrease in lattice parameter.

V_4C_3 is observed at large parameter values in the steels containing V, but has not been detected in association with the secondary hardening, although it must undoubtedly be present at that stage, but probably in so fine a dispersion as to be missed either due to the extracting technique or its very fine particle size. This situation is aggravated even further at lower (0·1%) vanadium contents when the V_4C_3 is not observed until the highest tempering parameters.

It has been observed that the carbides formed when a bainitic steel is tempered are Fe_3C and later $Fe_{21}Mo_2C_6$, but if the same steels were water-quenched, the carbides would be Fe_3C and finally Mo_2C. This change in behaviour may originate from segregation of the alloy element on a microscale in the bainitic steels or from the fact the Fe_3C is already present in the bainitic condition, and this may alter the nucleation conditions. This effect is being further investigated by considering the effect of prior structure on the tempering characteristics, and also the effect of other and larger alloy additions.

(*a*) 400° C, 50h, hardness 269 (*b*) 550° C, 4h, hardness 256 (*c*) 600° C, 10 h, hardness 229



(*d*) 600° C, 100h, hardness 208 (*e*) 700° C, hardness 155 (*f*) 700° C, 50 h, hardness 140

Fig. 16—Typical microstructures during tempering of a 1½% Mn, ½% Mo–B steel × 750

(c) Effect of Tempering on Mechanical Properties

It is to be expected that the general mechanical properties will follow a trend similar to the hardness values on tempering. Figure 18 shows the effect of tempering a typical bainitic steel containing 0·85% Mn and ½% Mo–B. The tensile strength is very little

(*a*) 400° C, 4 h, hardness 329 (*b*) 500° C, 10 h, hardness 336 (*c*) 600° C, 7 h, hardness 358

(*d*) 650° C, 7 h, hardness 324 (*e*) 700° C, 100 h, hardness 194 (*f*) 750° C, 50 h, hardness 177

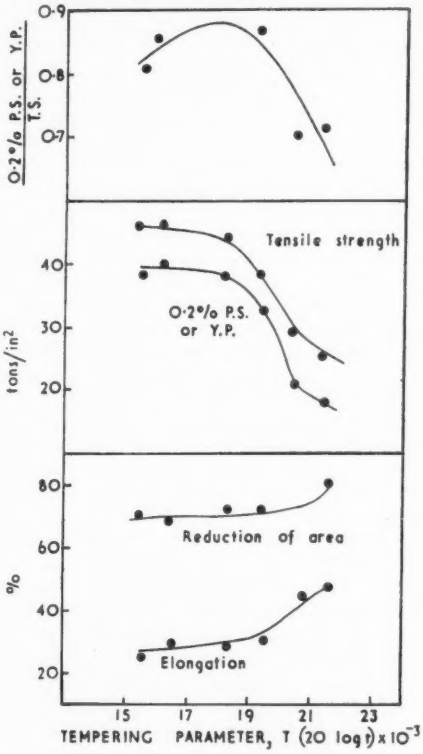
Fig. 17—Typical microstructures during tempering of a 1% Cr, 0·75% Mn, 0·3% V, 0·5% Mo–B steel × 750

affected until a tempering parameter of 18,000 is reached, after which the strength is reduced quite considerably. One interesting feature of the results is the ratio of 0·2% P.S. (or yield point) to tensile strength. It has already been explained that the proof stress ratio may be increased by tempering at 400° C. Figure 18 shows that this effect increases at higher tempering temperatures until values of 0·85–0·9 are obtained. This maximum is reached just before the major softening occurs, and the proof stress ratio then decreases again. This effect was noticed in all the tempering work on bainitic steels.

APPLICATIONS

The most obvious application for the type of steel described in this paper is as a high-strength weldable steel, since it is possible to obtain high strength in a low-carbon steel without heat-treatment and with a relatively high transformation temperature. When the requirements for a high-strength weldable steel are considered it will be seen that the possibilities are limited to three basic structures: ferrite and pearlite, bainite, and martensite.

Considering first the more general case of steels having a ferrite and pearlite structure, the steel can be strengthened in two ways. Firstly, the ferrite can be strengthened by small amounts of alloying elements in solid solution (the elements which are generally used being P, Si, Mn, Ni, Cr, or Cu). This has the added advantage that these elements also increase



Analysis: 0·14% C, 0·85% Mn, ½% Mo–B

Fig. 18—Effect of tempering on mechanical properties of a bainitic steel

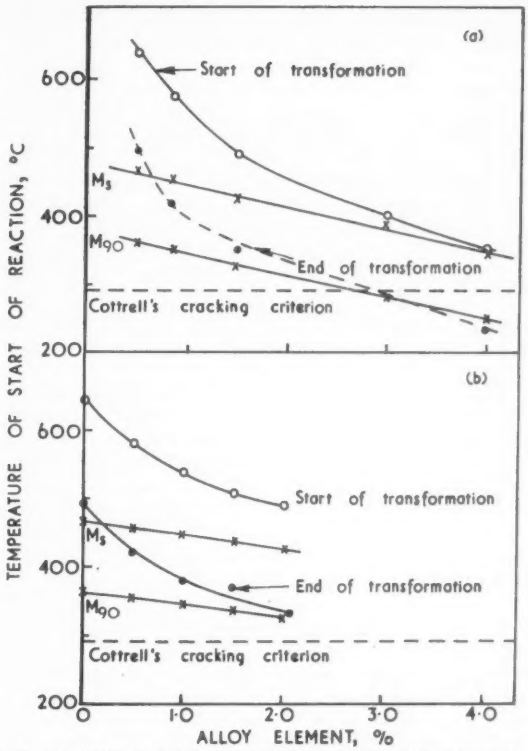


Fig. 19—Effects of (*a*) Mn and (*b*) Cr upon bainite and martensite transformation

the corrosion resistance of the steel. Alternatively, the steel can be strengthened by increasing the proportion of pearlite to ferrite. This is most effectively carried out by increasing the carbon content, but is also produced by the general effect of the alloying elements in changing the eutectoid composition and slowing down the transformation rate. An accompanying effect is that the lamellar spacing of the pearlite becomes finer, which also increases the strengthening effect.

The application of these principles led to the early development of 1½% Si, 1½% Mn, and 2–3% Ni steels. Of these, the 1½% Mn steel has continued and this general composition is standard in a number of countries. In addition during the last 20 years, there have been many weldable steels developed in different countries containing small amounts of alloying elements from the group Ni, Cr, Cu, P, Si, and V.

The limitations of this type of steel become apparent when the factors governing weldability are considered. During the welding operation, a local area of metal is heated into the austenitic region and is then quenched by the rapid conduction of heat away to the cooler regions of the plate. Consequently, the major factor to be considered is the response of the steel to quenching at a rate equivalent to the fastest quenching rate in the region adjacent to the weld.

This cooling rate is usually sufficient to produce martensite and it is the stresses associated with the formation of martensite which produce weld-cracking. Two factors are important, viz. the amount of stress

produced, and the temperature at which the transformation occurs. Carbon is harmful from both points of view. The transformation stresses increase with increasing carbon content and the transformation temperature is decreased. Alloying elements in general act to depress the transformation temperature. Work carried out by Cottrell[15] has shown that provided the transformation does not continue below 290° C no hard-zone cracking will occur. With the ferrite–pearlite structure both the carbon and alloy contents must be limited so that the transformation can remain above 290° C and this also limits the tensile strength. Even adopting the maximum strengthening effect of carbon and alloying elements it is doubtful whether properties much above 40 tons/in² can be obtained.

Once the limitations of a ferrite–pearlite structure have been accepted, there are only two alternatives. A low-carbon martensite is a possibility, since with a low carbon content the transformation temperature is reasonably high and the transformation stresses are low. The objection to this type of structure is that a quenching and tempering operation is necessary, which leads to distortion in thin sections and it is difficult to quench thick sections successfully.

The most suitable structure for a high-strength weldable steel is a low-carbon bainitic structure. As with the martensitic steel mentioned above, the low carbon content means that low transformation stresses are produced and, in any case, both the bainitic and martensitic transformations occur at a relatively high temperature. There is no difficulty with large sections because it has been shown earlier that properties can be obtained in up to 15-in. sections. The relationship between the transformation in these steels and Cottrell's cracking criterion is shown in Fig. 19. These diagrams relate to bainitic steels with varying manganese and chromium additions. The base composition is ½% Mo–B at the top level of carbon suggested for these steels, namely 0·15%. In each of these diagrams the lines representing the start and end of transformation are given, together with calculated M_s and M_{90} lines and the line at 290° C which represents Cottrell's cracking criterion.

The transformation lines indicated are for air-cooled 1-in. dia. bar, and these cooling rates might apply where pre-heating is used during welding. They do not, however, indicate the transformation range under the most rapid cooling conditions experienced during welding, since these cooling rates will certainly produce a martensitic transformation. The M_{90} line is of more interest for these conditions. Figure 19 shows that it is not until high alloy contents are reached that either the end-of-transformation line for 1-in. dia. bar, or the M_{90} line, crosses the Cottrell cracking line at 290° C. Using Mn or Cr, it should be possible to reach a tensile strength of 70 tons/in² (3% Mn or 2½% Cr addition) without conflicting with the Cottrell criterion.

Other applications which appear suitable are those where consistent properties are required in large sections. The high tensile strength and good tempering resistance makes them attractive for large rotor forgings, whilst the low carbon content and resistance to thermal cracking suggests suitability for die blocks.

DISCUSSION OF RESULTS

Some discussion has been given as the results were presented but it is now possible to summarize the factors controlling transformations of low-carbon steels in the bainitic range. The first essential feature is the retardation of high-temperature ferrite formation. Provided that the normal ferrite and pearlite transformations can be retarded, a wide range of cooling rates will produce a bainitic structure. The first question of importance, therefore, is how the alloying elements described (in particular ½% Mo–B) retard the normal ferrite formation.

Most of the normal alloying elements retard transformation by their effects on the diffusions which are necessarily associated with those transformations, and they retard both the pro-eutectoid and the bainitic ferrite reactions.

Boron, however, primarily affects transformations nucleating at austenite grain boundaries and therefore largely acts to retard the normal polygonal ferrite formation. In combination with a low carbon content, the required condition of a greatly retarded pro-eutectoid ferrite reaction and a bainite reaction only slightly retarded is achieved.

It is a combination of boron with another element which is most effective and in this respect the ½% Mo–B combination is most effective, with 1% W–B as the next most effective base. It is presumably the interaction of these two elements on the requisite diffusions which is responsible for their effect. Boron-free steels using a combination of alloying elements, but including Mo, can achieve bainitic structures at air-cooling rates, but the bainitic hardenability is low and the amount of alloy element necessary is higher than when boron is present. There appears to be no element that is as effective as boron in retarding ferrite nucleation at austenite grain boundaries.

Once a bainitic structure has been achieved it is then only necessary to depress the transformation temperature to raise the tensile strength. All the normal alloying elements can produce this effect, although some have a greater effect than others. The order of increasing effect would appear to be Cu, W, Ni, Mn, Cr, Mo, V, and Ti.

It has previously been suggested that the main reason for the increasing strength, with decreasing transformation temperature, is a reduction in grain size of the structure produced. The microstructures shown in Figs. 5*a–l* clearly indicate this effect. However, when the results for a large number of steels are examined it is found that some of the various alloying elements have a very similar effect in depressing the bainitic transformation temperature. An alloying element of 1% depresses the transformation temperature about 80° C. Consequently, the depression of transformation temperature and finer grain size produced can account for a strengthening of the structure but not for the variable effect of the different alloying elements. A further contribution would seem to be due to a solid-solution hardening effect of the alloying element, and the order of strengthening quoted above does agree to some extent with the known solid-solution hardening effect of the various alloying elements.

Once the two features mentioned above, viz. retardation of polygonal ferrite formation and depression of bainitic transformation, are understood, the general picture of low-carbon bainitic steels can be seen quite clearly. In fact it is felt that these steels present a simpler picture than any other type of steel and from the data contained in this paper it is possible to select a steel whose properties meet many requirements.

In designing for alloy steels using quenched and tempered steels it is usual to assume that providing efficient quenching is obtained, the strength depends on the carbon content. The alloying elements are effective mainly in controlling martensitic hardenability and tempering characteristics.

With the high carbon contents used (0·3–0·4%), the effect of carbon largely outweighs any effect due to the alloying elements.

With the bainitic steels described in this paper the lower carbon content allows each of the alloying elements to be effective and their contribution to the strength of a composition can be calculated from the formula given. The boron content ensures adequate bainitic hardenability and so, provided that a minimum base of ½% Mo–B is used, with carbon above 0·08%, there is no need to use the general alloy content to control hardenability. These elements can therefore be used to maximum efficiency in controlling strength.

There is a possible choice between alloying elements depending on their relative effect in strengthening and depressing the transformation temperature. For the best welding characteristics it is desirable to have the highest transformation temperature associated with any selected tensile-strength level. For this application therefore, those elements with the greatest solid-solution hardening effect would be incorporated. If the best impact properties are required, however, it is essential to have as low a transformation temperature as possible associated with a particular tensile-strength level to produce the finest grain size; consequently, alloying elements with lower solution-hardening effects would be used. Finally, if V is used, it is also possible to obtain very much improved tempering resistance.

The general conclusion is, therefore, that low-carbon bainitic steels offer a very attractive combination of mechanical properties which can be obtained in as-rolled or air-cooled sections. As the hardenability is high these properties can be obtained in large sections and the steels also have good welding properties and tempering characteristics. They should have many applications as high-strength weldable steels, forgings and die blocks.

SUMMARY

The main features of the low-carbon bainitic steels described in this paper are as follows:

(1) It is first necessary to retard the normal polygonal ferrite formation. This is achieved most successfully with a ½% Mo–B base composition, and can also be obtained with 1% W–B. It is also possible to use a boron-free higher alloy composition containing at least 0·3% Mo.

(2) Having retarded the polygonal ferrite formation, the steel transforms to bainitic ferrite over a wide range of cooling rates. This bainitic ferrite has superior mechanical properties to polygonal ferrite, mainly because of a much finer grain size. The grain size is controlled by the transformation temperature.

(3) The transformation temperature can be lowered by the addition of alloying elements, and a general relationship exists between transformation temperature and tensile strength. The strengthening effect of an alloying element is due partly to the depression of transformation temperature, and also to a solid-solution hardening effect. Within the bainitic transformation range it is possible to obtain tensile strengths between 40 and 80 tons/in².

(4) With larger alloy additions (e.g. 2% Cr), the bainitic transformation line is depressed and flattened. This means that the bainitic hardenability is high and consistent mechanical properties can be obtained in large section sizes (e.g. up to 24 in.).

(5) Bainitic steels have a characteristic tempering curve. Since the tensile strength depends upon the fine grain size, hardness is retained until ferrite grain growth occurs. Vanadium is particularly effective in providing good tempering resistance, and with 0·3% V some tempering resistance is obtained at temperatures up to 750° C.

(6) Owing to the low carbon content and the high transformation temperature, bainitic steels have excellent welding characteristics. Consequently, the most appropriate application is as a high-strength weldable steel.

Acknowledgments

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Sinter-bed Ignition

By H. Bates, B.Sc.

SYNOPSIS

A study has been made of the variables affecting sinter-bed ignition and hence the strength of the sinter from the top layer of the bed. The results show that there is an optimum rate of heat input and percentage of excess air for ignition and that an increase in the moisture content or the particle size of the top of the bed results in a decrease in the sinter strength. A total heat input of 4000 B.t.u./ft² was found to be sufficient for both the lean home and rich foreign-ore mixes used. 1473

Introduction

IN THE SINTERING of iron ores it is usually the case that the topmost layers of the sinter cake are weaker and more friable than the main body. It has long been recognized that this layer in fact contributes relatively little to the net output of blast-furnace sinter. Explanations have been put forward based on the fact that the attainment of an adequate sintering temperature is dependent largely on preheating the air for combustion by the hot layers of sinter above the flame front.[1] In the initial stages of the process this layer is non-existent and the surface of the bed must be sintered mainly by the heat supplied during ignition. The solid fuel in the topmost layers is liable, therefore, to burn out at a lower temperature level than in the deeper parts of the bed. Attempts to overcome this difficulty have included the provision of a thin layer of additional fuel on the bed surface and the development of the technique of double-layer sintering, in which the upper layer is prepared with a higher fuel content than the lower. Experimentally, attempts have been made to restrain the initial combustion of the solid fuel, until higher bed temperatures have been attained, by maintaining an inert atmosphere throughout the ignition period.

The view has also been put forward that the weakness of the surface layers is in part a consequence of the chilling effect of cold air drawn through immediately following ignition and it has been suggested that preheated air at this stage would lessen the thermal shock.

The present work has been carried out to study the effect of variations in the ignition conditions when using a single hood for ignition, notably the rate and total amount of heat input, the calorific value of the gas used, and the proportion of excess air drawn into the bed during ignition. This last variable is relevant to the inhibition of solid fuel combustion referred to above. Although the tests were carried out using two sinter mixes of widely different physical and chemical properties, the same experimental procedure has been adhered to throughout and, in order to focus attention on the processes in the top layer, the bed height was limited to 2 in.

EXPERIMENTAL PROCEDURE

In order that the investigation should be as complete as possible, the tests were carried out on two basic sinter mixes of widely different properties: a typical, lean, home-ore mix comprising Lincolnshire and Northamptonshire ironstones in the ratio 2 : 1 and a rich, foreign-ore mix comprising Mediterranean ore, Swedish ore, and Sierra Leone concentrates in the ratio 9 : 5 : 6. The return fines and coke contents of the dry, home-ore mix were 45% and 6%, respectively, and of the foreign-ore mix 30% and 6%, respectively. Throughout the tests the ores and return fines were screened to $< \frac{1}{4}$ in. and the coke to $< \frac{1}{8}$ in. except for specific tests to determine the effect of the materials' particle size on ignition. All materials were hand mixed on the floor and loaded into a special sinter box, area 4 ft², to a bed height of 2 in.

During ignition the gas and combustion air flows to the ignition hood and the volume of waste products drawn through the bed was measured and controlled. After ignition the airflow through the bed was kept constant at 75 ft³/min (S.T.P.*) per ft² of bed area up to and past the end of sintering until the bed of finished sinter was cold, thus simulating the cooling experienced by the sinter in the top layer of a normal deep bed. The figure of 75 ft³/min was chosen as being appropriate for a deep bed of raw mix during the first few minutes of sintering. At the end of the test the box was tipped on its side and the finished sinter cake, now a vertical slice, was prised out and allowed to fall flat. The shattered sinter was then loaded on to a $\frac{1}{2}$-in. sieve. This was placed in a mechanical screener and sieved for 3 min and the percentage $> \frac{1}{2}$-in. fraction determined. As most of the tests did not provide sufficient $> \frac{1}{2}$-in. sinter for a normal B.I.S.R.A. drop test to give reliable results, the percentage of $> \frac{1}{2}$-in. material after screening was adopted as a measure of the sinter strength. It was considered that, as the human element in the screening had been eliminated, the figures obtained would be suitable for comparison. They were, in fact, satisfactorily reproducible. It should be pointed out that due to the severity of this test a value of 50% represents a strong and well bonded sinter top.

A total heat input of 3000 B.t.u./ft² of bed was used as a standard except when investigating the effect of total heat input. The experiments were carried out using for ignition either normal coke-oven gas (calorific value 467 B.t.u./ft³ at S.T.P.) or simulated mixed coke-oven and blast-furnace gas (calorific value 150 B.t.u./ft³ at S.T.P.). The latter was obtained by diluting the normal coke-oven gas with a measured quantity of cooled inert combustion

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The author, who was at the North-East Coast Laboratories of the Ironmaking Division of B.I.S.R.A. when this work was done, is now with Huntington, Heberlein and Co., Ltd.

* S.T.P. = 60°F and 30 in. Hg.

products, i.e. $N_2 + CO_2$. In all cases the air supplied to the ignition burner was that required for complete combustion of the gas.

HOME-ORE SINTER-MIX IGNITION

Variables Examined

The following variables were examined during the series of tests to determine their effect on the strength of the top layer of sinter.

(i) Ignition intensity (B.t.u./ft²/min)
(ii) Percentage excess air drawn into the bed during ignition
(iii) Total heat input for ignition (B.t.u./ft²)
(iv) Material particle size
(v) Percentage moisture in the raw mix.

Ignition Intensity

The first set of experiments was carried out under both inert and oxidizing ignition conditions on the basic mix (moisture content 10%) to determine the optimum intensity for ignition when burning either coke-oven gas or a simulated mixed gas in the igniter. Table I shows the range of ignition intensities examined and the corresponding ignition times for a total heat of 3000 B.t.u./ft² of bed.

The results are shown in Fig. 1 from which it will be seen that the optimum ignition intensity when burning coke-oven gas is approximately 4500 B.t.u./ft²/min and with the simulated mixed gas 3000 B.t.u./ft²/min. It would appear that when using coke-oven gas the optimum ignition intensity increases slightly with increasing percentage excess air, whereas with the gas of lower calorific value the optimum intensity decreases with increasing percentage excess air.

Percentage of Excess Air drawn into Bed during Ignition

If during ignition the volume of combustion products from the ignition hood is balanced by the volume of waste gases drawn through the bed, then no excess or free air will be drawn into the bed, and the coke in the top of the bed is prevented from burning during the ignition period, a situation which may be termed 'inert ignition.' If the volume of waste gases is in excess of the volume of combustion products, free air will be drawn into the bed and the coke in the top of the bed will commence to burn during ignition, giving what may be termed 'oxidizing ignition.' The amount of air drawn through a sinter bed depends on the suction under the bed and the bed permeability. In tests carried out on the experimental sinter box, the airflow through the bed can

Table I
IGNITION INTENSITY

Ignition Intensity, B.t.u./ft²/min	Ignition Time for Total Heat of 3000 B.t.u. ft², s
1500	120
2250	80
3000	60
3750	48
4500	40
5250	34
6000	30

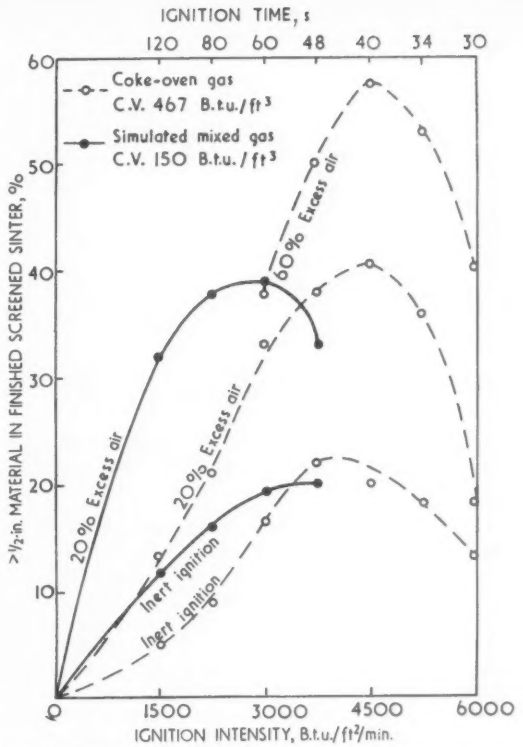
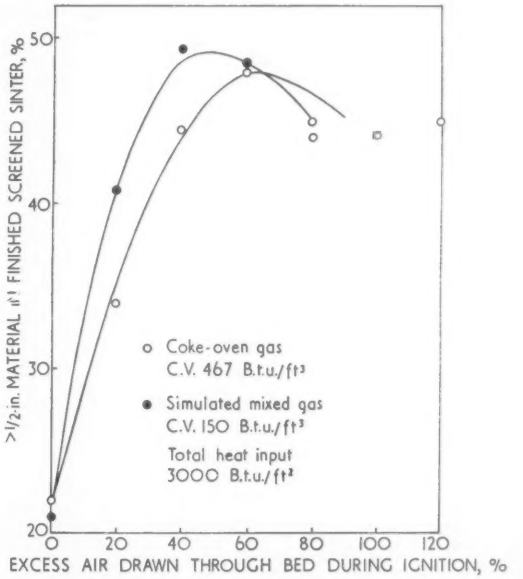


Fig. 1—Effect of rate of heat input (ignition intensity) on the strength of the sinter from the top two inches of a home-ore sinter-bed



Note.—**The point with abscissa and ordinate of approximately 80 and 45 respectively at the end of the curve for simulated mixed gas should be represented by a solid circle.**

Fig. 2—Effect of percentage excess air on the strength of the sinter from the top two inches of a home-ore sinter-bed

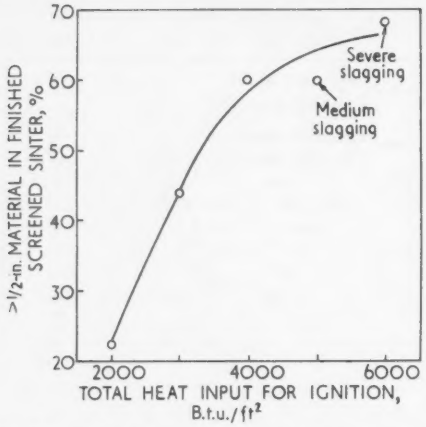


Fig. 3—Effect of total heat input on the top-layer sinter strength with a home-ore mix

be controlled, hence the effect on sinter strength of varying the amount of air drawn through the bed during the ignition period can be accurately determined.

Tests were carried out on the basic mix using a constant ignition intensity and total heat input, and varying the amount of excess air drawn into the bed. The total heat input used was 3000 B.t.u./ft² and the ignition intensity was 3750 B.t.u./ft²/min for coke-oven gas and 3000 B.t.u./ft²/min for the simulated mixed gas. The amount of excess air drawn into the bed during the ignition period has been calculated as a percentage of the theoretical air volume supplied to the hood to burn the ignition gas. Figures 1 and 2 show that a marked increase in sinter strength can be achieved, when using either gas, by drawing excess air into the bed during the ignition period. The optimum percentage of excess air when using coke-oven gas was found to be 60%; with the simulated mixed gas the optimum value was a little lower at 40%. There is an indication that, while quite high values make little difference to the performance with coke-oven gas, an increase beyond the optimum for the lean gas mixture may lead to progressive deterioration.

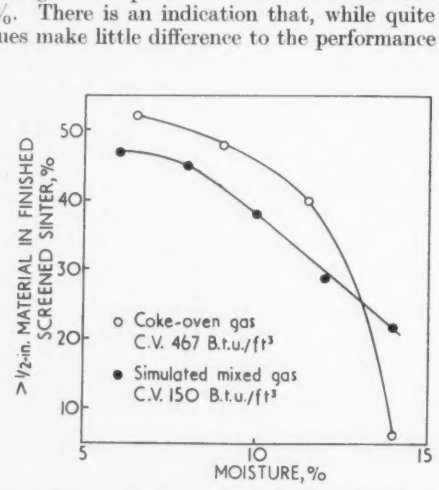


Fig. 4—Effect of moisture content of a home-ore raw mix on the top-layer sinter strength

Total Heat Input

The effect of total heat input on the strength of the top-layer sinter was investigated using coke-oven gas for ignition. A constant intensity of 3750 B.t.u./ft²/min together with 60% excess air was used and the total heat input varied from 2000–6000 B.t.u./ft². The results are shown in Fig. 3. With heat inputs of up to 4000 B.t.u./ft² there was no indication of surface slagging and the sinter strength increased fairly rapidly with increasing total heat input. Higher total heat inputs resulted in severe slagging of the bed with little further increase in sinter strength.

Material Particle Size

Reducing the coke size from $< \frac{1}{8}$-in. to $< \frac{1}{16}$-in. had no noticeable effect on the strength of the top-layer sinter. This may well have been due to the fact that the $< \frac{1}{8}$-in. coke used as a standard contained 70% of $\frac{1}{16}$-in. size.

Tests were then carried out using the basic mix composition and constant ignition conditions, but varying the particle size of the ores and return fines. The results are shown in Table II.

From Table II it will be seen that there is no marked difference in the strength of the sinter from the top of the bed when using either the $< \frac{1}{4}$-in. or the $< \frac{1}{8}$-in. materials, but that the presence of $\frac{1}{2}$–$\frac{1}{4}$-in. material results in a decided decrease in strength. Suggested reasons for this decrease in strength are:

(i) The presence of large particles in the bed surface results in uneven and patchy ignition and hence a uniform flame front is not established until lower down the bed
(ii) The large particles, having a greater heat capacity, take longer to attain the temperature necessary for agglomeration and may not reach it during the ignition period
(iii) The surface contact between particles is lower at larger sizes; hence the formation of a strong bond between them is harder to achieve.

Percentage Moisture in the Raw Mix

Tests were carried out on the basic mix to determine the direct influence of the percentage moisture content of the raw mix on sinter ignition, and hence on the strength of the sinter in the top layer of the bed. A total heat input of 3000 B.t.u./ft² was used for ignition. The results are plotted in Fig. 4 which shows clearly that a decrease in the mix moisture content is

Table II
EFFECT OF PARTICLE SIZE ON SINTER STRENGTH

Ore and Return Fines Size Analysis				Finished Sinter Strength, % > ½-in. after screen test
Nominal Size, in.	½–¼ in.	¼–⅛ in.	⅛ in.	
$< \frac{1}{2}$	17%	45%	38%	17·0
$< \frac{1}{4}$	...	55%	45%	48·0
$< \frac{1}{8}$	...	...	100%	52·0

accompanied by a progressive increase in the strength of the sinter from the top of the bed. However, the percentage of moisture in the raw mix not only affects the sinter ignition, as shown by Fig. 4, but also the pre-ignition and average sintering permeabilities and the segregation of the raw mix on the strand. Table III shows the segregation of coke in the bed obtained at varying moisture contents for a nominal $< \frac{1}{2}$-in. mix, comprising 49% Lincs-Northants ore, 45% return fines, and 6% coke, and Table IV the sintering characteristics of the mix at the different moisture contents.

Good coke and size segregation is desirable since large lumps of ore, which have a deleterious effect on the strength of the top layer sinter as shown by Table II, are practically eliminated from the top of the bed and the concentration of coke is increased, both of which should enhance the chances of getting sinter of sufficient strength from this region of the bed.

From Table IV it is clear that, although an increase in the moisture content of the mix results in a large increase in the pre-ignition permeability, its effect on the average sintering permeability and the flame front speed is not so marked, and that, at the lower moistures, the advantages gained by the improved segregation and ignition are becoming apparent as shown by the quality of the finished sinter cake.

FOREIGN-ORE SINTER-MIX IGNITION

Variables Examined

(i) Ignition intensity (B.t.u./ft²/min)
(ii) Percentage excess air drawn into the bed during ignition
(iii) Total heat input for ignition (B.t.u./ft²).

Ignition Intensity

The first set of experiments was carried out under oxidizing ignition conditions to determine the optimum ignition intensity when burning coke-oven gas in the igniter. Total heat inputs of 3000 and 5000 B.t.u./ft² were used for ignition and 60% excess air was drawn into the bed during the ignition period. The ignition intensity was varied in stages from 3000 to 6000 B.t.u./ft²/min, giving ignition times of from 60 to 30 s for a total heat input of 3000 B.t.u./ft², and from 100 to 50 s for a heat input of 5000 B.t.u./ft². In both cases the optimum ignition intensity was approximately 4500 B.t.u./ft²/min, the peaks being less pronounced than for the home ore mixes.

Percentage of Excess Air

Tests were carried out on the basic mix (moisture 7%) using a total heat input of 3000 B.t.u./ft² for ignition at an intensity of 4500 B.t.u./ft²/min and the excess air drawn into the bed was varied from 0·0 to 100%. The amount of excess air drawn into the bed during the ignition period has been calculated as a percentage of the theoretical air volume required to burn the ignition gas. The results show that the strength of the top layer sinter increases with increasing excess air up to 60% excess air; higher values, however, result in a progressive deterioration in sinter strength (Fig. 5).

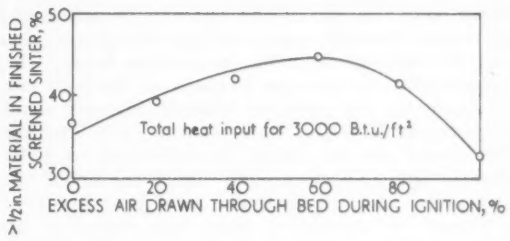


Fig. 5—Effect of percentage excess air on the strength of the sinter from the top two inches of the bed with a foreign-ore mix

Total Heat Input

The effect of total heat input on the strength of the sinter from the top of the bed was investigated using a constant ignition intensity of 4500 B.t.u./ft²/min and 60% excess air, and varying the total heat input from 3000 to 6000 B.t.u./ft² of bed. Very little increase in sinter strength is gained by increasing the total heat input above 4000 B.t.u./ft². With a total heat input of 6000 B.t.u./ft², slight slagging of the bed surface was becoming apparent and this tended to reduce the permeability of the bed.

CONCLUSIONS

In general, the results of the tests with the foreign ore sinter mix compare very closely with those obtained with the home-ore mix.

The experimental results show that for a given mix an appreciable improvement in the quality of the

Table III

COKE SEGREGATION IN A 10-in. BED

Distance from bed top, in.	% Coke		
	8% Moisture	10% Moisture	12% Moisture
0–2	7·92	8·20	5·80
2–4	7·16	7·12	7·00
4–6	7·08	6·92	6·00
6–8	6·64	6·60	6·10
8–10	1·70	1·70	5·51
Average % coke	6·08	6·09	6·07

With mix moistures of 8% and 10% the top of the bed was predominantly $< \frac{1}{2}$-in. material

Table IV

SINTERING CHARACTERISTICS IN A 10-in. BED

	8% Moisture	10% Moisture	12% Moisture
Pre-ignition permeability, B.P.U.	93	130	150
Average sintering permeability, B.P.U.	93	108	110
Average flame front speed, in./min	0·93	0·99	1·00
% Finished sinter cake $> \frac{1}{2}$-in.	59	58	44
Strength index of $> \frac{1}{2}$-in. sinter	67	62	54

sinter from the top of the bed can be achieved if particular attention is given to the rate of heat input and the atmosphere under the ignition hood. They suggest that, for the gas mixtures as used on most sinter plants, the optimum ignition intensity should be between 3500 and 4000 B.t.u./ft²/min, and that the strand suction should be controlled to give a slightly oxidizing atmosphere under the hood. For any given mix the optimum total heat input will probably be dependent upon the chemical and physical properties of the constituents, but for the mixes used in the tests there does not appear to be any advantage in exceeding 4000 B.t.u./ft² of bed provided the top of the bed is predominantly $< \frac{1}{4}$-in. material, as higher values result in partial slagging of the surface with very little increase in sinter strength. Particles above this size have been shown to have a deleterious effect on the strength of the sinter from the top of the bed. The moisture content of the raw mix is perhaps the most important of the operating variables studied during the tests and is probably the easiest to vary on a plant. It affects not only the sinter ignition but also the permeability and segregation of the mix, a high moisture content resulting in a high pre-ignition permeability but in poor ignition and segregation. It has been shown that at a lower moisture content, although the pre-ignition permeability is lower, the average sintering time and permeability are not appreciably altered, and that the advantages gained from the improved segregation and ignition result in an all round increase in the blast-furnace sinter output and strength.

It is interesting to note that over the range of ignition conditions examined, when using the foreign-ore mix, there is less variation in the strength of the sinter from the top of the bed than that obtained over the same range of conditions when using the home-ore mix. This would seem to suggest that ignition conditions are much less critical when operating a foreign-ore practice. A possible explanation is that a larger proportion of the heat supplied during the ignition period is utilized in evaporating the additional moisture and dissociating the carbonates usually associated with typical home-ore mixes. Thus much more depends on the efficient utilization of the heat in the ignition gases and the coke in the top of the bed than when using foreign ores, and any deviation from the optimum conditions is likely to result in a rapid deterioration in the strength of the sinter from the top of the bed.

Finally, although cold air was drawn into the bed immediately after ignition, a strong top layer sinter was obtained in every case, provided that ignition was carried out under the optimum conditions. It must be borne in mind, however, that the optimum ignition conditions specified in this paper might not be applicable when using preheated air after ignition.

It is also appreciated that individual plant conditions may require some variation from the figures put forward in this paper in order to compensate for radiation and cooling water losses, which are a function of the design of the ignition equipment.

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The Tempering of Low-alloy Creep-resistant Steels containing Chromium, Molybdenum, and Vanadium

By E. Smith, M.A., Ph.D., and J. Nutting, M.A., B.Sc., Ph.D.

SYNOPSIS

The hardness changes which occur during the tempering of a range of quenched 0·2% C steels containing up to 3% Cr, 1% V, and 1% Mo have been measured. The accompanying microstructural changes were determined by preparing extraction replicas and examining these with the electron microscope; the precipitated carbides were identified by transmission electron diffraction and selected-area electron diffraction.

The results indicate that vanadium carbide is nucleated separately within the ferrite grains as small plates, which grow larger with further tempering. Chromium stabilizes ϵ-iron carbide, which then transforms to cementite and finally to Cr_7C_3. The rate of growth of chromium carbide is much greater than that of vanadium carbide. The addition of molybdenum modifies the form of the precipitated cementite, but Mo_2C is not precipitated until high tempering temperatures are attained. In the presence of chromium and vanadium, molybdenum carbides are not readily formed.

During tempering the ferrite matrix recrystallizes, but, if large carbide particles are formed at the ferrite grain boundaries, ferrite grain growth may be hindered.

The results obtained are discussed in relation to the previously established creep properties of ferritic low-alloy steels. 1472

Introduction

WHEN PLAIN CARBON STEELS are tempered from the martensitic condition, within the temperature range 400–700° C, their hardness decreases logarithmically with tempering time. If the steels contain non-carbide-forming alloying elements or small amounts of chromium this general type of softening behaviour still prevails, although the rates of softening are decreased. With the addition of strong carbide-forming elements such as molybdenum

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Table I
ANALYSES OF STEELS USED

Cast No.	Type	C	Si	Mn	Cr	Mo	V	V/C
SG 8556	0·7% V	0·22	0·59	0·59	...	...	0·73	3·32
SG 8558	3% Cr	0·23	0·55	0·49	2·96	...	...	...
SG 8555	0·5% Mo	0·21	0·54	0·57	...	0·55	...	...
SG 8557	0·7% V–0·5% Mo	0·21	0·58	0·61	...	0·54	0·72	3·43
SG 8559	3% Cr–0·5% Mo	0·22	0·45	0·49	3·02	0·54	...	...
SG 8560	3% Cr–0·7% V	0·22	0·52	0·51	3·08	...	0·73	3·32
SG 8561	3% Cr–0·7% V–0·5% Mo	0·22	0·47	0·52	3·18	0·50	0·75	3·40
SG 8490	1% Cr–Mo–V	0·22	0·23	0·46	1·03	0·99	0·20	0·91
SG 8491	,,	0·22	0·24	0·46	1·05	0·99	0·39	1·77
SG 8492	,,	0·23	0·26	0·49	1·09	0·94	0·59	2·48
SG 8493	,,	0·23	0·28	0·52	1·14	0·98	0·73	3·18
SG 8494	,,	0·22	0·28	0·54	1·10	0·96	0·87	3·96
SG 8495	,,	0·22	0·28	0·53	1·14	0·97	1·11	5·05

or vanadium the softening behaviour on tempering is greatly modified, and the hardness/tempering-time relationships at different temperatures show characteristics identical with those of an age-hardening alloy; a maximum hardness is reached, the value of which decreases with increasing tempering temperature. This phenomenon in alloy steels has become known as secondary hardening and has been widely investigated.

As a result of the investigations of many workers, it has been found that secondary hardening is associated with the formation of alloy carbides. The techniques used for the investigations have involved measuring changes of magnetic properties, electrical resistivity, internal friction, and the thermal, dilatometric, and specific volume changes which occur on tempering. More direct evidence of the structural changes has been obtained by extracting the carbide phases from the tempered alloys and identifying these by chemical analysis, by X-ray, and electron diffraction examination to obtain the crystal structure, and by electron microscopical examination of the dispersed residues. It has, however, until recently been impossible to follow directly the changes in microstructure which occur during secondary hardening because they were on too fine a scale to be observed with the optical microscope and the specimen-preparation techniques were not sufficiently advanced to allow the electron microscope to be used to the limit of its capabilities. But now, as a result of the development of the direct carbon-extraction replica method by the present authors,[1] it has become possible to make full use of the electron microscope in following the microstructural changes associated with the tempering of steels.

The present investigation was carried out to determine the microstructural changes occurring during the tempering of a series of 0·2% C steels containing up to 3% Cr, 1% V, and 1% Mo. Steels containing various combinations of the three alloying elements were made and heat-treated, and their hardness determined. The microstructures were examined with the aid of the electron microscope and an attempt was made to identify the precipitated carbides by electron diffraction examination of the carbides on the extraction replicas. The chromium–molybdenum–vanadium steels are widely used because of their good creep resistance, and so an attempt has been made to correlate the observed microstructures with previously established creep data for these steels.

Materials Used and Heat-treatments

The steels were made by the United Steel Companies Ltd. as 18-lb melts in a high-frequency induction furnace starting with an aluminium-killed base material. They were cast into 3-in. moulds and then forged into bars of $\frac{3}{4}$ in. diameter. The analyses are given in Table I.

The specimens for heat-treatment were taken from the $\frac{3}{4}$-in. bar by cutting slices $\frac{3}{8}$ in. thick. These were then heated at 1050° or 1080° C for one hour and quenched in oil. The tempering was carried out at 400°, 500°, and 600° C in a lead bath and at 650°, 675°, 700°, and 750° C in air. The tempering times were $\frac{1}{2}$, 1, 2, 5, 10, 20, 50, and 100 h at each temperature. After tempering the specimens were air cooled.

Hardness Measurements and Metallographic Technique

After heat-treatment the specimens were sectioned through a diameter and parallel to the bar axis and the exposed surfaces used for hardness measurements

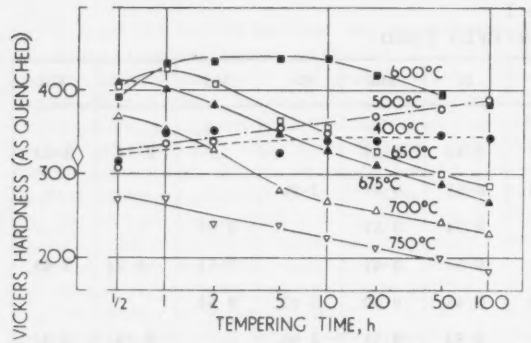


Fig. 1—Effect of isothermal tempering upon the Vickers hardness of the 0·7% V steel

Note—**In Figs. 1, 8, 12, 16, 18, and 19, the vertical axis represents the Vickers hardness of the specimens after the tempering treatments shown on the horizontal axis. The hardness of the steel in the as-quenched condition is marked with a diamond symbol on the vertical axis.**

and metallographic examination. The Vickers hardness values at a load of 30 kg were obtained from the mean of measurements on five indentations. The surfaces for metallographic examination were first abraded, then electro-polished in a 20% perchloric–acetic acid bath and finally mechanically polished with alumina cream. All the specimens were etched in nital, the concentration of which was varied from 1% for steels containing no chromium to about 5% for the 3% Cr steels. The etching time was 10–15 s.

To prepare the replicas, a carbon film about 200 Å thick was deposited on the polished metal surface using the technique described by Bradley.[2] The film was then stripped from the metal by electro-polishing and caught on a mounting grid. This gave a direct carbon replica. Another method of stripping the carbon film was to mark it into squares and then immerse the specimen and film in a suitable etching reagent. This dissolved the ferrite and, when the film floated free, the carbide particles present in the surface of the steel were attached to the film. This gave a direct carbon extraction replica. Fuller experimental details have been given by the authors.[1]

The replicas were examined in the Siemens electron microscope (UM60C). Transmission electron diffraction patterns were obtained both in the electron microscope and in a " Finch " type Edwards electron diffraction camera.

EXPERIMENTAL RESULTS

Steels containing a Single Alloying Element

0·2% C–0·7% V Steel

The secondary hardening of the vanadium steel is strongly marked (Fig. 1). Although it is customary to illustrate the effects of temperature and tempering time on alloy steels by plotting the influence of temperature at constant tempering time upon the hardness, the tendency of a system to show precipitation hardening is more clearly illustrated if the hardness values are plotted against the logarithm of the tempering time for each tempering temperature. This convention has been adopted. On tempering at 400° C little change of hardness occurred, but at 500° C the hardness started to increase above that of the original martensite and was still rising after 100 h at this temperature. After tempering at 600° C, a maximum in the hardness was found which was about 100 D.P.N. above that of the original martensite. With higher tempering temperatures, the maximum hardness attained was lower and occurred after shorter tempering times.

Houdremont *et al.*[3] have shown that in steels containing vanadium the secondary hardening is associated with the formation of the alloy carbide V_4C_3. From the examination of carbide extracts, Crafts and Lamont[4] concluded that the vanadium carbide grew out of a lacy cloud-like structure. Koch and Wiester[5] found that with the onset of secondary hardening, i.e. tempering between 500° and 550° C, the vanadium content of the carbide residue increased from 9% to 67%; they were also able to show that the size of carbide particles formed at 600° C was about 200 Å. Kuo,[6] using a technique similar to that of Crafts and Lamont, obtained evidence of a thread-like form of V_4C_3 at the onset of secondary hardening. The threads were 30 Å dia. and 200 Å long. At higher temperatures the particles were thought to be spheroids of about 1000 Å dia.

Microstructural changes on tempering the martensite—From the present investigation it has been found that after tempering the microstructures could be divided into three groups: those showing iron carbide only, those with iron carbide and vanadium carbide, and those showing vanadium carbide only. After tempering for up to 100 h at 400° C the only carbide present was Fe_3C which was identified by its electron diffraction pattern. A typical micrograph is shown in Fig. 2. The carbide was initially feathery in shape but became more rounded with increasing tempering time. With up to 20 h tempering at 500° C the structures were similar to those found at 400° C,

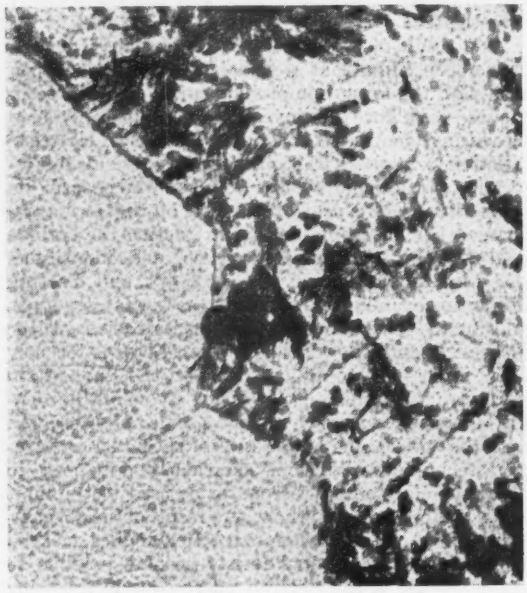


Fig. 2—0·7% V steel tempered for 100 h at 400° C showing cementite. Carbon extraction replica × 25,000

but after 50 h treatment the structure within the ferrite grains began to change and the electron diffraction patterns from the extraction replicas showed rings characteristic of V_4C_3, but weak. The principal ring was that of the (220) reflection at a spacing of 1·46 Å. The ring for (200) planes of vanadium carbide was masked by rings from the iron carbide. The structure after ½ h at 600° C was similar to that found after 100 h at 500° C and showed both V_4C_3 and Fe_3C. Treatment for 5 h at 600° C resulted in a striking change from the ½ h specimen; most of the iron carbide had dissolved and had been replaced by vanadium carbide. The electron diffraction pattern derived from the vanadium carbide is shown in Fig. 3*a*. The rings are continuous and broad which suggests that they are produced by a large number of very small particles; furthermore, certain of the rings are absent or weak. These correspond with the reflections from the planes of three indices, i.e. the (h, k, l) planes. The rings from the planes (h, 0, 0) and (h, k, 0) are present. This suggests that the V_4C_3 is in the form of thin plates. The microstructure (Fig. 4) shows a cloud-like distribution of very small spots, some considerably less than 100 Å dia. These spots represent the vanadium carbide particles at an early stage of their precipitation from the ferrite.

It is possible to estimate roughly the size of the vanadium carbide plates from measurements on the electron diffraction photographs using the formula given by Pinsker,[7]

$$t = \frac{L\lambda}{B}$$

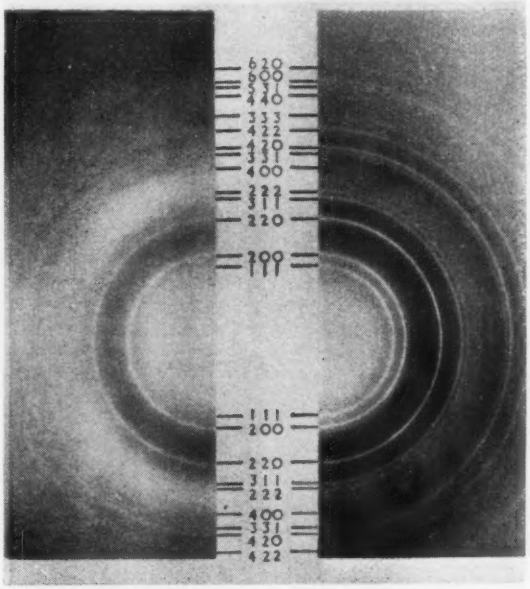
where t = thickness of specimen in a direction of a given reflection, L = specimen–plate distance, λ = wavelength of radiation, and B = half-peak width of the diffraction ring on the plate. In the case of electron diffraction the value $L\lambda$ can be readily obtained since $L\lambda = dr$, where d = lattice spacing of reflecting planes and r = radius of diffraction ring on the plate. Measurements of half-peak widths were made with a microdensitometer on the pattern from the specimen tempered for 5 h at 600° C, which corresponds to peak hardness. For the (110), (200), and (220) rings the thickness of the diffracting crystals in the various crystallographic directions were then calculated as follows:

[111]	40Å
[200]	60Å
[220]	60Å

It was then possible to calculate the axial dimensions of the particles, which were: a = 60 Å, b = 40–60 Å, and c = 25 Å. The accuracy of these values is about 10–20%, the least accurate being the value for the c axis.

The shape of the vanadium carbide particles corresponds therefore to flattened ellipsoids.

After tempering for 100 h at 600° C the electron diffraction rings were still diffuse, but there were spots within the rings. The microstructure showed that the particles within the grains were still very small but at the ferrite grain boundaries preferential growth had occurred, and this had resulted in denudation of the fine precipitate in the areas adjacent to the grain boundary. The grain-boundary precipitates account for the spots on the diffraction pattern. Individual particles of V_4C_3 were first observed after tempering for 100 h at 650° C, whilst after 100 h at 675° C the carbides were clearly visible and were present as a Widmanstätten precipitate within the ferrite grains (Fig. 5). Most of the particles are



(*a*) 0·7% V steel tempered for 5 h at 600° C (*b*) Standard pattern from 0·7% V steel tempered for 100 h at 700° C

Fig. 3—Electron diffraction patterns of vanadium carbide

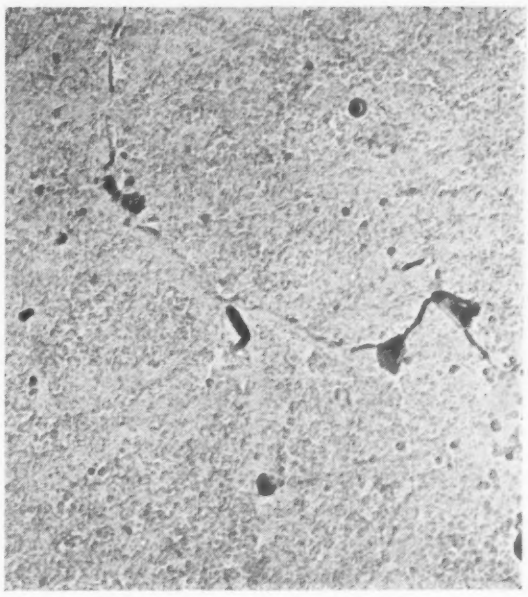


Fig. 4—0·7% V steel tempered for 5 h at 600° C showing cloud-like precipitation of vanadium carbide. Carbon extraction replica × 25,000

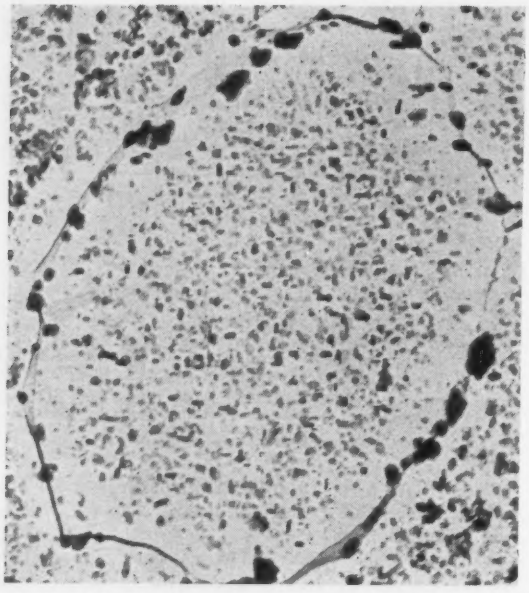


Fig. 5—0·7% V steel tempered for 100 h at 675° C showing Widmanstätten precipitation of vanadium carbide within ferrite grains. Carbon extraction replica × 25,000

plate-like but some appear as needles. It may be assumed that the latter represent plates viewed from the edge. Tempering for 100 h at 700° C gave much larger particles of V_4C_3, about 400 Å dia., but there were still some small particles of 100 Å dia. After treatment at 750° C the microstructures showed evidence that the specimens had been in the austenitic region, but the vanadium carbide particles were still small.

The diffraction patterns, obtained from the extraction replicas of the specimens tempered in the range 650–700° C, showed a progressive sharpening of the rings, particularly of those corresponding to the planes of three indices. The pattern found after 100 h at 700° C is shown in Fig. 3*b*. But even after this treatment the relative intensities of the diffraction rings did not correspond to those found in X-ray powder photographs of V_4C_3. It is assumed that this is because the particles on the extraction replica retain their plate-like shape.

Tempering of retained austenite—In the as-quenched specimen, regions were found around the austenite grain boundaries, occupying 5–10% of the total area, in which the structure was unlike that in the normal martensitic regions. Under polarized light the areas extinguished at all angles, which suggested that the material was isotropic. Reflection electron diffraction from the surface of the specimen showed the presence of austenite. It was assumed therefore that these areas represented regions of retained austenite. On tempering, these regions were still visible after 100 h at 675° C but after treatment at 700° C they became difficult to distinguish. Selected-area electron diffraction patterns were obtained from extraction replicas taken from the specimens after tempering. These showed the presence of vanadium carbide in the austenitic areas even after ½ h at 400° C, but the rings were very broad, indicating a small particle size. With increasing tempering time and temperature, the diffraction rings became sharper but their relative intensities remained the same. This suggests that when vanadium carbide is precipitated from austenite it tends to form as spheres rather than plates. The electron micrographs showed the presence of small particles of 50 Å dia. at the initial stage of the precipitation.

Recrystallization and grain growth in the ferrite matrix—During the tempering of plain carbon steels Hyam and Nutting[8] have suggested that at temperatures of 500° C the ferrite matrix recrystallizes, and that subsequent grain growth is controlled by the growth of the cementite particles which are present almost entirely at the ferrite grain boundaries. From the present investigation evidence of matrix recrystallization has also been obtained.

By examining the replicas at a low magnification (× 1500) it has been possible to observe changes in the ferrite matrix. At low temperatures (400–500° C) the ferrite grains, when visible, were plate-like and probably corresponded in size with those of the original martensite. After treatment for 100 h at 600° C the microstructure showed equiaxed grains of 3–10μ dia. At the grain boundaries there were small particles of dissolving Fe_3C and growing V_4C_3, whilst within the grains there was a fine precipitate of V_4C_3 (*see* Fig. 6). The specimen tempered for 100 h at 700° C showed a different structure (Fig. 7). The ferrite grains were less than 3μ in dia. and were plate-like with large particles of V_4C_3 at their grain boundaries. It seems difficult to believe that the structure found after the 700° C treatment was

Fig. 6—0·7% V steel tempered for 100 h at 600° C showing equiaxed ferrite grains. Carbon extraction replica × 3000

obtained via the structure found after the 600° C treatment.

To investigate the phenomenon further a specimen was given a double temper, i.e. 100 h at 600° C followed by 10 h at 700° C. In this case, equiaxed ferrite grains were found. The following mechanism is suggested to explain the development of these microstructures. On tempering at 600° C, the feathery Fe_3C particles in the boundaries of the ferrite (earlier martensite) grains begin to dissolve as V_4C_3 is precipitated. Eventually nearly all the Fe_3C dissolves and the structure consists of an almost uniform dispersion of V_4C_3. Under these conditions recrystallization and grain growth may occur and the latter proceeds by boundary movement from one set of particles to the next. Diffusion of vanadium along the grain boundaries to form large particles is slow at 600° C; therefore the grains are able to grow until they are stabilized by undissolved Fe_3C or any large V_4C_3 particles. Once this happens, the V_4C_3 particles in the boundaries grow larger and thus hinder further grain growth. At 700° C vanadium diffusion is rapid even within the ferrite grains, whilst vanadium carbide can be precipitated immediately at the grain boundaries without the prior precipitation and subsequent replacement of Fe_3C. It is the precipitates of vanadium carbide at the acicular ferrite grain boundaries, which then hinder further grain growth. The microstructures obtained after tempering directly for 100 h at 650° and 675° C confirm these views in that they show some equiaxed grains and some acicular grains.

0·2% C–3% Cr Steel

Bain[9] has shown that steels containing less than 12% Cr as the only carbide-forming element do not show appreciable secondary hardening. The hardness results obtained in the present investigation (given in Fig. 8) do not show any tendency for secondary hardening to occur, but there is a low rate of softening particularly at the tempering temperatures of 400° and 500° C. In the temperature range 600–750° C the change in hardness with tempering time is very similar to that of a plain carbon steel.

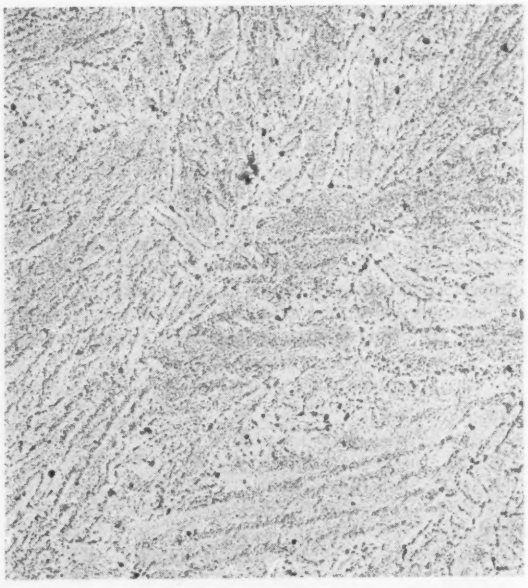


Fig. 7—0·7% V steel tempered for 100 h at 700° C showing plate-like ferrite grains. Carbon extraction replica ×3000

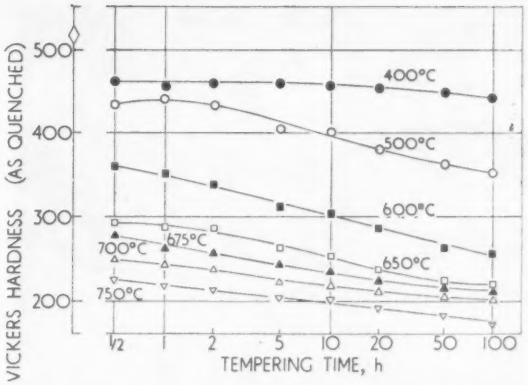


Fig. 8—Effect of isothermal tempering at 400°, 500°, 600°, 650°, 675°, 700°, and 750° C upon the Vickers hardness of the 3% Cr steel

The nature of the alloy carbides formed on tempering steels containing chromium has been fully investigated. Balluffi, Cohen, and Averbach[10] have shown that, with a 4% Cr steel, Cr_7C_3 is formed, and they have obtained some evidence which indicates that the alloy carbide is nucleated directly from the ferrite. Crafts and Lamont[11] believe that the alloy carbide is formed from the initially precipitated iron carbide, the change taking place through a transitional lattice. A somewhat different view has been put forward by Wever and Koch.[12] They have obtained evidence to show that Cr_7C_3 is nucleated at the cementite/ferrite interface and that the Cr_7C_3 then grows at the expense of the cementite.

Kuo[13] and Goldschmidt[14] showed that prior to the formation of Cr_7C_3 the cementite was enriched with chromium to the extent of 20%. When the Cr_7C_3 was nucleated, both chromium and carbon were available in the cementite for the growing particle and therefore growth could take place quickly. The later results of Kuo[6] were not conclusive in confirming the mechanism of the $Fe_3C \rightarrow Cr_7C_3$ phase change he proposed.

Microstructural changes on tempering—In the present investigation it was found that after tempering at 400° C for ½ h the microstructure showed acicular particles within the ferrite grains and sheet-like precipitates at the grain boundaries. By selected-area electron diffraction it was found that the intragranular precipitates gave a few rings characteristic of ϵ-iron carbide, whilst the pattern from the intergranular precipitates corresponded to Fe_3C. Further tempering for up to 100 h produced little change in the microstructure. Tempering at 500° C for up to 100 h produced only slight changes in the microstructures from those found at 400° C. A typical

Fig. 9—3% Cr steel tempered for 100 h at 500° C showing Widmanstätten and grain boundary precipitates of cementite. Carbon extraction replica × 20,000

micrograph is given in Fig. 9. There was, however, a change in the diffraction patterns; no rings corresponding to ϵ-iron carbide could be identified and the only carbide present was cementite. It may be concluded therefore, in the absence of microstructural change, that the ϵ-iron carbide changes directly to cementite without the latter being separately nucleated. Within the ferrite matrix small spots were observed but it is believed that these represent etch markings as no electron diffraction pattern could be obtained from them. Similar markings have been observed by van Wijk[15] in pure iron and by Coiley[16] in plain carbon steels. It is thought that the markings are small humps produced by a slower rate of etching at dislocations surrounded by carbon atmospheres. There were about 10^{10} of these etch markings per cm^2, a density similar to that observed by the other workers.

Fig. 10—3% Cr steel tempered for 100 h at 600° C showing rod-like Widmanstätten precipitates of chromium carbide (Cr_7C_3). Carbon extraction replica × 12,500

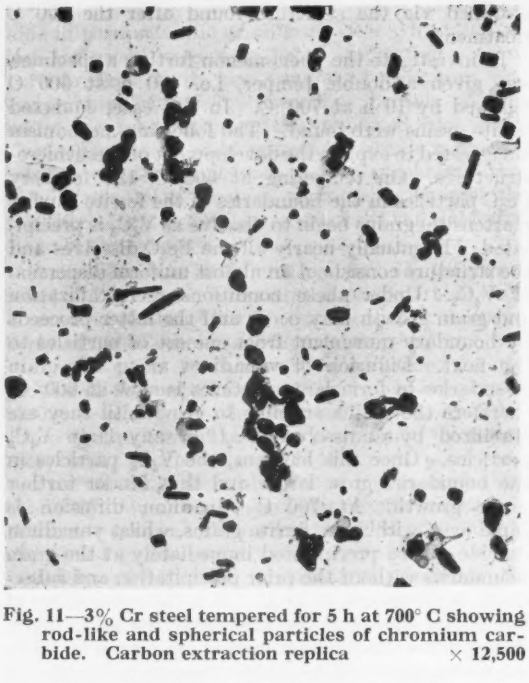


Fig. 11—3% Cr steel tempered for 5 h at 700° C showing rod-like and spherical particles of chromium carbide. Carbon extraction replica × 12,500

On tempering at 600° C pronounced changes occurred. The structure after ½ h was identical to that found after 100 h at 500° C and the diffraction pattern showed only rings of cementite. With longer treatment times the carbides at the grain boundaries spheroidized, whilst those within the grains assumed a distinctive rod-like shape. This change of microstructure corresponded with the appearance in the diffraction patterns of rings characteristic of Cr_7C_3. The structure obtained after 100 h at 600° C is shown in Fig. 10. There was no evidence of the micro-

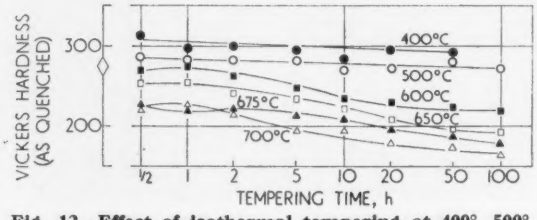


Fig. 12—Effect of isothermal tempering at 400°, 500°, 600°, 650°, 675°, and 700° C upon the Vickers hardness of the 0·5% Mo steel

structural changes characteristic of the precipitation of vanadium carbide. It may be concluded therefore that the Cr_7C_3 forms from a transformation within or close to the cementite particles as postulated by Kuo.[13]

Tempering within the temperature range 650–750° C resulted in microstructures in which the rod-like and spherical particles became larger in size and fewer in number, as shown in Fig. 11. There was no metallographic evidence of retained austenite in the steel.

Recrystallization and grain growth of the ferrite—In the latter stages of tempering the ferrite grains were not clearly visible, since they were not outlined with precipitated carbides. It was difficult therefore to determine the conditions of tempering under which recrystallization and grain growth occurred. Recrystallization appeared to begin at 600° C when the grain-boundary carbides had spheroidized, but even after 100 h at 750° C, when considerable grain growth had occurred, the grains were still elongated.

0·2% C–0·5% Mo Steel

Bain[9] and Houdremont *et al.*[3] have shown that steels containing sufficient molybdenum will secondary harden, but that molybdenum is not as efficient as vanadium in bringing about this effect. In the present investigation it was found that with only 0·5% Mo in the steel there was little tendency to secondary harden. The results obtained at different tempering temperatures and times are shown in Fig. 12. There was retarded softening after tempering at 400° and 500° C, but at 600°, 650°, and 675° C there was a logarithmic decrease in hardness with tempering time. Rapid softening occurred after 10 h at 700° C. The A_1 transformation temperature was exceeded by treatment at 750° C and consequently the results were not considered further.

Fig. 13—0·5% Mo steel tempered for 5 h at 500° C showing plate-like precipitates of cementite. Carbon extraction replica × 12,500

Fig. 14—0·5% Mo steel tempered for ½ h at 675° C showing branch-like precipitates of cementite. Carbon extraction replica × 12,500

The secondary hardening of the steels containing 2–5% Mo has been shown by Kuo[17] to be associated with the formation of Mo_2C. By examining extracted carbides he found that the Mo_2C was precipitated between 500° and 600° C in the form of needles of 30 Å dia. and 200 Å long. Kuo suggested that needles were nucleated separately within the ferrite matrix but Wever and Schrader[18] have concluded that the Mo_2C could form in or from cementite. In low-molybdenum steels Kuo[13] has found a double carbide which he designates M_aC_b, but from the results of Smith quoted by Kuo[13] the double carbide only forms after long tempering times at 700° C.

Microstructural changes on tempering—After treatment at 400° C feathery precipitates of Fe_3C were formed at the boundaries of plate-like ferrite grains, with heavy precipitation at the pre-existing austenite grain boundaries. The amount of precipitated carbide increased with tempering time. Tempering treatments of up to 10 h at 500° C gave structures similar to those found at 400° C. A micrograph is given in Fig. 13. But at longer tempering times the carbide particles changed their shape to a branched form. It is thought that this is a type of spheroidization rather than renucleation and growth of a new carbide. The electron diffraction patterns from the branched carbides gave clearly defined rings characteristic of Fe_3C. Tempering at 600° C and above resulted in a rapid change from the feathery to the branched form of the carbide of the type shown in Fig. 14. After 100 h at 650° C needle or blade-like precipitates were first observed within the ferrite grains. These were very similar in shape to the precipitates of Mo_2C found by Kuo[17] in 2–5% Mo steels.

The needle-shaped precipitates were also observed after 50 h at 675° C and after 2 h at 700° C. In most cases they occurred individually within the ferrite grains, but sometimes they were associated with the cementite particles as shown in Fig. 15. On tempering longer at 700° C the cementite particles became fewer and larger whilst the needle-shaped particles became more like short sticks and were difficult to distinguish from the elongated cementite particles. No carbides of the M_aC_b type were detected by electron diffraction.

Recrystallization and grain growth—There appears to be some recrystallization at 600° C corresponding with the change from the feathery to the branched form of the Fe_3C, and this is also associated with a fall in hardness. With continued treatment there was little grain growth until 675° C when the branched carbide started to spheroidize further. Equiaxed gains were then formed with large carbide particles at their grain boundaries.

Steels containing Two Alloying Elements

There are three steels in this group. The results obtained will be described by considering the effect of molybdenum upon the steels containing vanadium and chromium respectively and then the effect of vanadium upon the steel containing chromium.

0·2% C–0·7% V–0·5% Mo Steel

The behaviour of this steel on tempering is very similar to that of the 0·2% C–0·7% V steel. The addition of 0·5% Mo brought about an increase of 36 in the Vickers hardness of the as-quenched steel when compared with that of the carbon–vanadium steel. After tempering at the various temperatures and times, it was found that the hardness values after any given time and temperature treatment were 20–40 D.P.N. above those of the carbon–vanadium steel. It may be concluded, therefore, that the effect of the 0·5% Mo is to act chiefly as a ferrite strengthener rather than contribute as a precipitation hardener. However, with the addition of molybdenum, peak hardness is reached at slightly shorter times at any given temperature. This effect would occur if the molybdenum increased the rate of diffusion of the vanadium in the ferrite.

Microstructural changes on tempering—The presence of 0·5% Mo in the steel prevented there being any retained austenite. It also raised the A_1 transformation temperature above 750° C.

The microstructural changes after treatment at temperatures up to 700° C were identical with those found in the 0·2% C–0·7% V steel. Tempering at 750° C still gave plate-like particles of V_4C_3 within the ferrite grains and large spherical particles at the grain boundaries. There was no evidence of molybdenum carbide in any of the specimens examined.

0·2% C–3·0% Cr–0·5% Mo Steel

The behaviour of this steel on tempering was very similar to that of the 0·2% C–3% Cr steel. It was found that tempering at 400° C gave almost identical results for the two steels. After tempering at 500° C the hardness of 0·2%C–3%Cr–0·5% Mo steel fell less rapidly than that of the 0·2% C–3% Cr steel. At all the higher tempering temperatures there was an almost logarithmic decrease in hardness with tempering time. The effect of the molybdenum was to raise the hardness above that of the carbon–chromium steel after a given treatment.

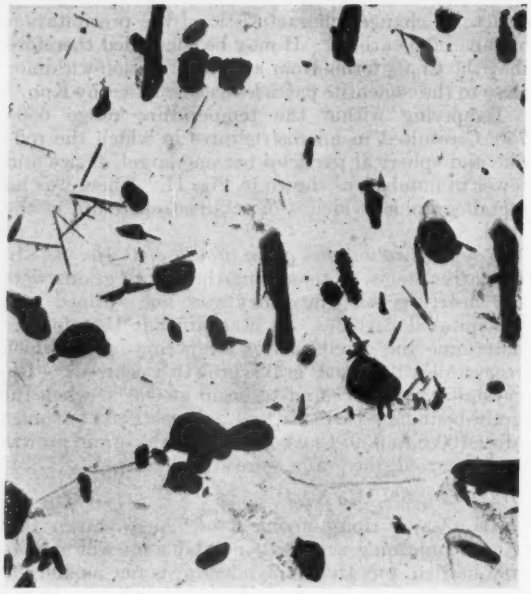


Fig. 15—0·5% Mo steel tempered for 5 h at 700° C showing needle-like precipitates of Mo_2C. Carbon extraction replica. × 12,500

Microstructural changes on tempering—After all the treatments up to 10 h at 500° C no differences could be observed between the structures of the chromium and chromium–molybdenum steels. However, after this treatment the diffraction pattern from the chromium–molybdenum steel showed a less distinct cementite pattern than the chromium steel. This result would be obtained if the effect of the molybdenum were to stabilize the ε-iron carbide, and this would also account for the slow rate of fall of the hardness after tempering at 500° C.

After tempering at 600° C, the ragged carbides at the ferrite grain boundaries did not spheroidize as rapidly as in the chromium steel and this effect was also observed after the higher-temperature treatments, but was less marked than at 600° C. There was no evidence of the formation of needle-shaped molybdenum carbides even after treatment for 100 h at 750° C.

It would appear from these results that the effect of the molybdenum in the chromium steel is to increase the general level of hardness and to retard the softening of the steel on tempering. This may be contrasted with the effect of molybdenum on the vanadium steel where the effect is to increase the general level of hardness but accelerate the hardening and softening of the steel on tempering.

0·2% C–3% Cr–0·7% V Steel

On tempering, this steel showed characteristics of both the chromium and the vanadium steels examined

previously. The hardness results obtained after the different tempering treatments are shown in Fig. 16. The steel did not secondary harden, as did the carbon–vanadium steel, but the softening on tempering occurred less readily than with the carbon–chromium steel.

Microstructural changes on tempering—Treatment for ½ h at 400° C resulted in the formation of a precipitate within plate-like ferrite grains, and a film-like precipitate at the grain boundaries. The electron diffraction pattern showed only a few rings, some of which corresponded to those from Fe_3C and the others to those from ε-iron carbide. With further tempering at 400° C the precipitates within the grains became less ragged and more acicular whilst the electron diffraction showed more rings characteristic of Fe_3C.

After tempering for 5 h at 500° C changes in the background of the ferrite grains were observed which suggested that precipitation of vanadium carbide was being initiated. At 600° C the vanadium carbide precipitated more rapidly than in the carbon–vanadium steel, and other large carbides were also present in the microstructure. Although it was not possible to obtain a diffraction pattern from these carbides, it may be assumed that they are Cr_7C_3. Shaw and Quarrell[19] and Ward and Rait[20] have shown that in steels of this composition chromium and vanadium carbides may co-exist.

Tempering in the range 650–750° C resulted in growth of both the V_4C_3 and Cr_7C_3 particles and a decrease in their number. A typical microstructure showing both carbides is given in Fig. 17.

Recrystallization and grain growth—Recrystallization and grain growth had started at 650° C but the grains were stabilized by the presence of Cr_7C_3 at their grain boundaries. After tempering at 675° and 700° C, recrystallization seems to have been stopped by the rapid precipitation of V_4C_3 at the boundaries of plate-like ferrite grains but the rapid growth of the V_4C_3 coupled with the fact that large particles of Fe_3C or Cr_7C_3 were always present in the microstructure made grain growth difficult at all temperatures. At no stage was there a microstructure similar

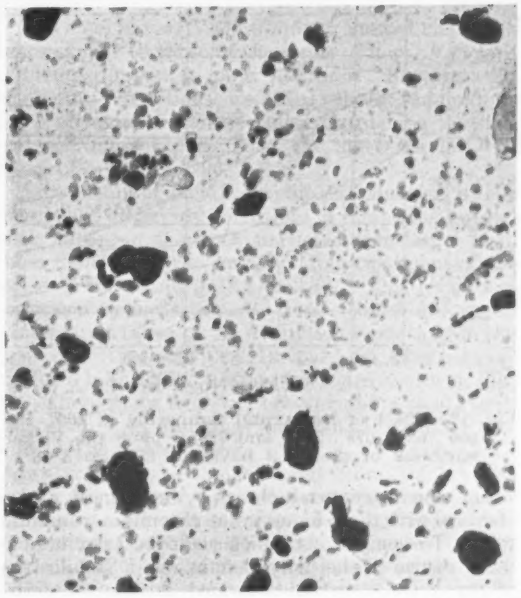


Fig. 17—3% Cr–0·7% V steel tempered for 5 h at 700° C showing large particles of Cr_7C_3 and small particles of V_4C_3. Carbon extraction replica × 25,000

to that found in the carbon–vanadium steel, where there was only a finely dispersed precipitate of V_4C_3 and no large particles to hold up movement of grain boundaries. Treatment at 750° C resulted in recrystallization and grain growth in spite of the heavy precipitation at the grain boundaries.

Steels containing Three Alloying Elements

0·2% C–3% Cr–0·7% V–0·5% Mo Steel

The behaviour of this steel on tempering was almost identical with that of the chromium–vanadium steel, but with a given treatment the hardness of the chromium–vanadium–molybdenum steel was slightly higher. A comparison of some of the results obtained with the two steels is shown in Fig. 18.

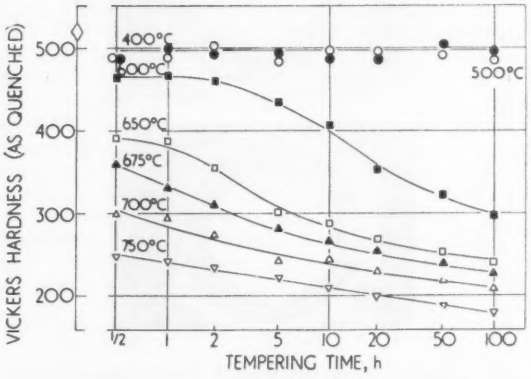


Fig. 16—Effect of isothermal tempering at 400°, 500°, 600°, 650°, 675°, 700°, and 750° C upon the Vickers hardness of the 3% Cr–0·7% V steel

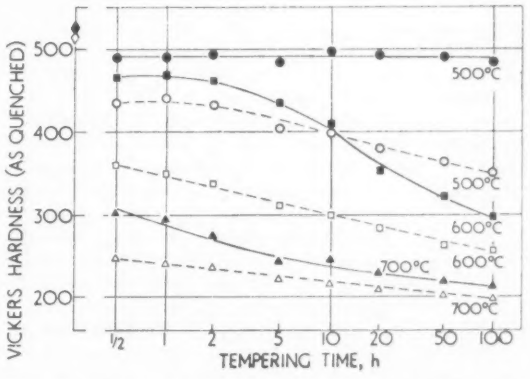


Fig. 18—Effect of isothermal tempering at 500°, 600°, and 700° C upon the Vickers hardness of the 3% Cr–0·7% V and 3% Cr–0·7% V–0·5% Mo steels

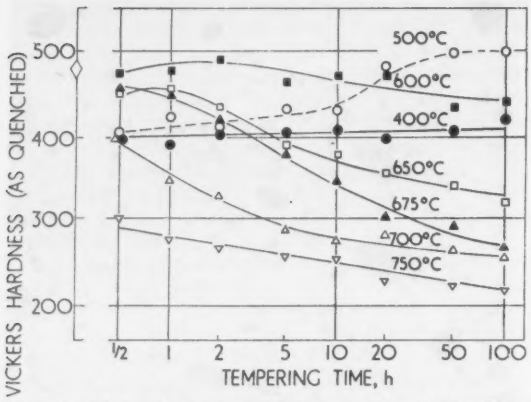


Fig. 19—Effect of isothermal tempering at 400°, 500°, 600°, 650°, 675°, 700°, and 750° C upon the Vickers hardness of the steel having V/C ratio 3·18

The microstructural changes were again almost identical with those found in the chromium–vanadium steel. The only effect which could be attributed to the presence of the molybdenum was a stabilization of the acicular carbides formed at low tempering temperatures; they persisted even after 5 h at 650° C.

Effect of Varying the Vanadium/Carbon Ratio in 0·2% C–1% Cr–1% Mo Steel

Other workers have shown that the creep properties of chromium–molybdenum–vanadium steels are greatly dependent upon the vanadium/carbon ratio. In view of this the hardness and metallographic changes which occurred on tempering a range of vanadium–steels were investigated. Six steels were examined having V/C ratios of 0·91, 1·77, 2·48, 3·18, 3·96, and 5·05. The results obtained for the steel having V/C ratio 3·18 are described first since this steel is of most importance commercially as it gives the best creep properties. The results obtained with the steels of lower and greater V/C ratios are then described with reference to the steel of V/C ratio 3·18.

This steel is also important in relation to the other steels described previously, since its composition is intermediate between that of the 0·7% V–0·5% Mo and the 3% Cr–0·7% V–0·5% Mo steels.

Vanadium content 0·73%, V/C ratio 3·18

The hardness changes on tempering are intermediate between those of vanadium–molybdenum and the 3% chromium–vanadium–molybdenum steels. The results are shown in Fig. 19. The hardness was still increasing after 100 h at 400° and 500° C, but a peak was reached after 2 h at 600° C and after ½ h at 650° C. An almost linear decrease of hardness with the logarithm of the tempering time was found after treatment at 675° and 700° C. In general peak hardness was obtained after a shorter time with this steel than with the vanadium–molybdenum steel.

Microstructural changes—After ½ h at 400° C the microstructure showed plate-like ferrite grains with sheet-like carbides at both the ferrite and prior austenite grain boundaries. The electron diffraction patterns were those of Fe_3C. Little change of structure was observed after increasing the tempering time up to 100 h.

Vanadium carbide precipitates were first observed after 5 h at 500° C and the sequence of changes with longer tempering times and higher temperatures was that of coarsening of the cloud-like precipitates and the growth of Cr_7C_3 particles.

Recrystallization and grain growth—Conditions for recrystallization and grain growth appear to be similar

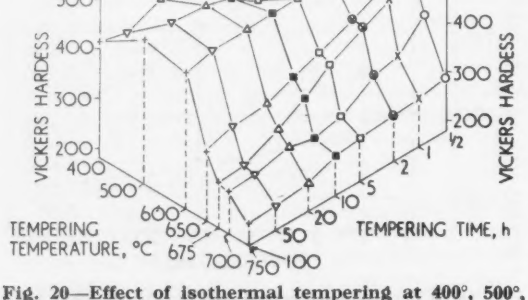
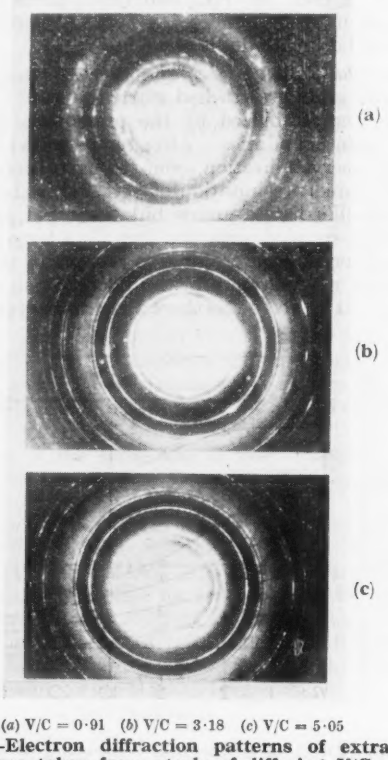


Fig. 20—Effect of isothermal tempering at 400°, 500°, 600°, 650°, 675°, 700°, and 750° C upon the Vickers hardness of the steel having V/C ratio 2·48



(*a*) V/C = 0·91 (*b*) V/C = 3·18 (*c*) V/C = 5·05

Fig. 21—Electron diffraction patterns of extraction replicas taken from steels of differing V/C ratios tempered for 5 h at 700° C

to those in the vanadium and vanadium–molybdenum steels. Equiaxed grains are formed by recrystallization at 600° C whilst at 700° C the rapid precipitation of vanadium carbide stabilizes the plate-like ferrite grains.

Vanadium content 0·2%, V/C ratio 0·91

The changes in hardness on tempering were similar to those found in the steel of V/C ratio 3·18, but the actual hardness values and degree of secondary hardening were lower. On tempering at 500° C peak hardness was reached after 50 h.

Microstructural changes—The initially formed iron carbides did not all dissolve as the vanadium carbide formed on tempering at temperatures up to 675° C. The vanadium carbides precipitated in dense patches within the ferrite matrix and this heterogeneity of the precipitate distribution may be partially responsible for the low creep-resistance of this steel. After 100 h at 700° C, $Cr_{23}C_6$ and Mo_2C were detected.

Vanadium contents 0·39% and 0·57%, V/C ratios 1·77 and 2·48

The hardness values obtained on tempering these two steels were intermediate between those of the steels having V/C ratios of 0·91 and 3·18. With increasing vanadium content the secondary hardening became more marked and peak hardness was reached after longer tempering times at a given temperature. With the steel having V/C ratio of 1·77 there was a retardation of the softening between 650° and 675° C and a further hardening peak was developed with the steel of V/C ratio 2·48. This retardation or increase in hardness disappeared as the time of tempering was increased. The effect is shown in Fig. 20.

The microstructural changes were similar to those found in the steel of V/C ratio of 0·91, but with increasing vanadium content there was a larger amount of V_4C_3 precipitated and a smaller amount of residual cementite. Also the precipitated V_4C_3 particles were more uniformly distributed within the ferrite matrix. The unusual changes in hardness between 650° and 675° C at short tempering times could not be associated with the formation of any new type of precipitate.

At 675° C the rate of dissolution of iron carbide is much greater than at 650° C. This can be inferred from the remaining iron carbide observed in the microstructure after corresponding times at the two temperatures. As a result, the volume fraction of V_4C_3 obtained after short times at 675° C is greater than that at 650° C. It is thought that this increased volume fraction of the finely dispersed carbide is responsible for the increase in hardness. At longer times of tempering the volume fractions of V_4C_3 will be almost identical at 650° C and 675° C and therefore the hardness is only a function of particle size. Thus lower hardness values were obtained after 5 h at 675° C than after 5 h at 650° C.

Vanadium contents 0·87% and 1·11%, V/C ratios 3·96 and 5·05

The behaviour of the steel having V/C ratio 3·96 was almost identical with that of the steel having V/C ratio 3·18. The results obtained will therefore not be described further.

The steel having V/C ratio 5·05 showed characteristics similar to those of the steels of lower V/C ratio on tempering up to peak hardness. But once the tempering conditions exceeded those required to bring about maximum secondary hardening the steel softened more rapidly than the steels having V/C ratios of 3·18 and 3·96.

As a result of metallographic examination it was found that under any conditions of tempering beyond peak hardness, the V_4C_3 particles were larger than those found after the corresponding treatment of the steels of V/C ratios 3·18 and 3·96. This result was confirmed by measurements from electron diffraction photographs taken of the extraction replicas from the steels of V/C ratios 0·91, 3·18, and 5·05, after tempering for 5 h at 700° C. The results are shown in Fig. 21. The pattern from the low-vanadium steel shows rings from V_4C_3 and rings and spots from other carbides. The pattern from steel of V/C ratio 3·18 shows very strong spotted rings from V_4C_3, arced regions from the V_4C_3 particles at the ferrite grain boundaries, and very few spots from other carbides. The pattern from the high-vanadium steel is derived only from V_4C_3 whilst the spots making up the rings are larger than those obtained in the pattern from the steel of V/C ratio 3·18.

DISCUSSION OF RESULTS

Precipitation of Vanadium Carbide from a 0·2% C–0·7% V Steel

The changes in hardness which occur on tempering for constant times at different temperatures are characteristic of a secondary hardening steel and of a normal precipitation hardening system.

It appears that the increase in hardness is associated with the precipitation of V_4C_3 in the ferrite matrix. However, the vanadium may have other effects at tempering temperatures below those necessary to bring about the precipitation of the carbide. The addition of 0·7% V to 0·2% C steel lowers the rate of softening at tempering temperatures up to 400–500° C. This may be a result of:

(i) Solid-solution hardening of the ferrite by the vanadium
(ii) The precipitation of iron carbide and its stabilization by vanadium
(iii) The formation of vanadium-enriched zones prior to the precipitation of vanadium carbide (cf. the formation of G.P.[1] zones in Al–4% Cu alloys).

Solid-solution hardening will certainly occur since there is a difference in size between the vanadium and iron ions. But in view of the slight increase in hardness from the as-quenched condition after tempering for 5 h at 500° C, it is thought that the vanadium must exert some other effect. Precipitation hardening may occur as a result of the formation of iron carbides, but in plain carbon steels this effect is found at tempering temperatures below 400° C. The vanadium however appears to slow down the rate of growth of iron carbide; this could be the result of two effects. Vanadium may stabilize the iron carbide by replacing iron in the cementite lattice, but, as the solubility of vanadium in cementite is low, the stabilizing effect is thought to be small.

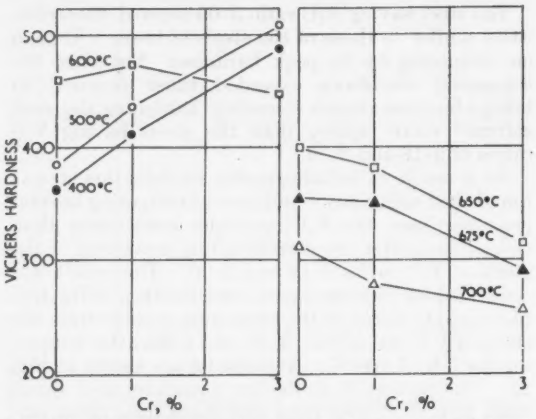


Fig. 22—Effect of chromium content upon the Vickers hardness of steel containing 0·7% V and 0·5–1·0% Mo after tempering for 10 h at 400°, 500°, 600°, 650°, 675°, and 700° C

Alternatively, the growth rate of the carbides may be lowered by vanadium lowering the rate of diffusion of carbon in the α-iron to such an extent that this becomes the rate controlling process rather than the self-diffusion of α-iron as suggested by Hyam and Nutting[8] for carbide growth in plain carbon steels. Le Claire[21] states that 1% W lowers the diffusion rate of carbon in α-iron, and it is probable therefore that vanadium, a stronger carbide-former than tungsten, may exert a similar influence. The mechanism by which this could occur would be an association between the vanadium and the iron atoms within the ferrite lattice, similar to that postulated by Fast[22] in the case of manganese and carbon in ferrite.

The fact that spheroidization of the precipitated iron carbide is taking place on tempering within the range 400–500° C whilst an increase of hardness is found suggests that some other precipitation reaction is being initiated.

At 400° C the 10% retained austenite decomposes directly to give vanadium carbide and this may be partly responsible for the increase in hardness. But the vanadium–molybdenum steel, in which no retained austenite is observed, shows similar characteristics although the rate of hardening is lower.

It may be that prior to the precipitation of V_4C_3 an ordering of the vanadium and carbon atoms takes place within the ferrite matrix, producing hardening similar to that obtained during the formation of G.P. [1] zones when ageing Al–4%Cu alloys. There is some microstructural evidence to support this view. After tempering at 400° C the ferrite matrix shows a structure of small spots, but these change to ridges immediately prior to the appearance of V_4C_3 rings in the diffraction pattern from the extraction replicas. More information on the early stages of the precipitation of vanadium carbide might be obtained if thin films of this steel could be examined directly in the electron microscope, as in the examination of aluminium alloys by Castaing.[23]

According to Richardson[24] V_4C_3 is more stable than cementite at 0° C and at an activity of vanadium of 1%. Yet on tempering, iron carbides are precipitated first, and V_4C_3 does not form until after 5 h at 500° C. The reason for this is undoubtedly the slow rate of diffusion of vanadium in ferrite.

The addition of 0·5% Mo to the 0·7% V steel increases the hardness generally and reduces slightly the time taken to reach peak hardness at temperatures above 500° C. As there is no evidence of precipitation of Mo_2C and as it is unlikely that molybdenum dissolves in vanadium carbide it seems that molybdenum will remain in solid solution in the ferrite.

If the precipitation of vanadium carbide is controlled by the rate of diffusion of vanadium the presence of the large molybdenum atoms in the ferrite should, by increasing the lattice strain, accelerate the diffusion of vanadium. This would result in the more rapid secondary hardening which is observed.

The 3% Cr Steel

The hardness determinations show that the chief effects of the chromium are to raise the martensitic hardness and the hardness after tempering at 400–500° C well above the corresponding values for a 0·2% C steel. The structural reasons for the high hardness at 400–500° C appear to be that Widmanstätten precipitation of ε-iron carbide which occurs at low tempering temperatures is stabilized by the presence of chromium. Recently Baker[25] has shown that the chromium replacement may take place to a considerable extent in ε-iron carbide. Also the chromium may lower the rate of diffusion of carbon in the ferrite. The ε-carbide maintains partial coherency with the ferrite matrix and spheroidizes only slowly. Therefore little softening occurs with increasing temperature until cementite starts to form. The change from ε-carbide to cementite has been shown by Baker to be associated with a decrease in chromium content of the extracted carbides. As the ε-carbide changes directly into cementite there are likely, therefore, to be high local concentrations of chromium in the region of the newly-formed cementite particles.

As the tempering temperature increases the chromium content of the carbides once again increases. This effect has been noted by many workers and is thought to be a result of the replacement by chromium of iron in the cementite lattice prior to the formation of Cr_7C_3. However, in the matrix adjacent to the newly formed cementite particles there will be a high chromium content and Cr_7C_3 may be nucleated in this region. If this is so the number of Cr_7C_3 particles which form is never likely to be much greater than the number of cementite particles to which they are adjacent. Therefore no increase in hardness is likely to occur during the transformation $Fe_3C \rightarrow Cr_7C$, i.e. the 3% Cr steel does not secondary-harden.

The addition of 0·5% Mo to a steel containing 3% Cr raises the general level of hardness and seems to stabilize further the ε-carbide. The stabilization could result from molybdenum replacement in the ε-carbide and by further lowering of the diffusion rate of the carbon in the ferrite. This latter effect may account for the lower rate of spheroidization of the cementite in the chromium–molybdenum steel than in the chromium steel.

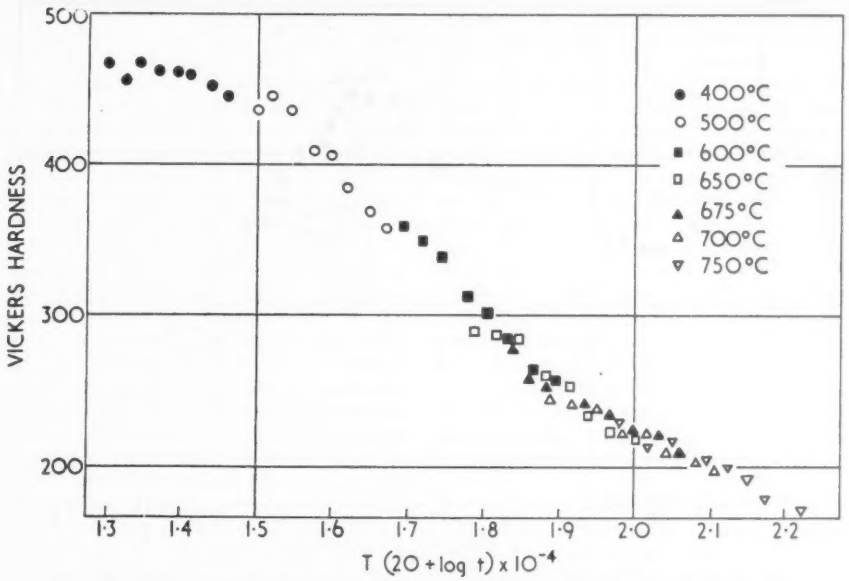


Fig. 23—Change of hardness with tempering parameter for the 3% Cr steel

Interaction of Chromium and Vanadium

The addition of 3% Cr greatly modifies the tempering behaviour of the 0·2% C–0·7% V steel. The hardness measurements indicate that the chromium–vanadium steel no longer secondary-hardens. (This may be inferred from Fig. 16 where none of the isothermal lines intersect.) This result may seem rather surprising in view of the metallographic evidence for the precipitation of V_4C_3.

By considering the results obtained with the 3% Cr steel in relation to those found with the 3% Cr–0·7% V steel, the absence of secondary hardening may be readily explained.

After tempering at 400° C, or for short times at 500° C, the high hardness appears to be maintained by the stabilization of the ϵ-carbide as with the 3% Cr steel, and this effect may be enhanced by the presence of vanadium. With longer times at 500° C and 600° C the softening due to the change from ϵ-carbide to cementite is counteracted by hardening due to precipitation of vanadium carbide.

As a result of these two mechanisms the hardness of the 3% Cr–0·7% V steel after tempering for up to 100 h at 400° C or 500° C and 2 h at 600° C is greater than the peak hardness at 600° C of the 0·7% V steel.

After longer times of tempering at 600° C or higher tempering temperatures, the 3% Cr–0·7% V steel is softer than the 0·7% V steel. The more rapid rate of softening in the more highly alloyed steel results from the formation of Cr_7C_3 at the expense of V_4C_3, and the increased rate of growth of the V_4C_3 in the steel containing chromium.

Richardson[24] has shown that, at low activities of vanadium and chromium, V_4C_3 has a higher free energy of formation than Cr_7C_3, but with increasing activity the free energy of Cr_7C_3 becomes the greater. The present results are in conformity with this view but it appears that V_4C_3 and Cr_7C_3 can co-exist over a wide range of tempering conditions. This has also been observed by Shaw and Quarrell.[19]

The addition of molybdenum appears to have little effect upon the microstructural changes in the chromium–vanadium steels. It is possible, therefore, by considering the steels containing 0·7% V–0·5% Mo, 3% Cr–0·7% V–0·5% Mo, and 1% Cr–0·7% V–1·0% Mo (V/C ratio 3·18) to determine the effects of increasing chromium content upon the hardness values after different tempering treatments. The results obtained after tempering for 10 h at various temperatures are shown in Fig. 22. These indicate that after tempering at 400° and 500° C the hardness increases with increasing chromium content. This is probably because the chromium is stabilizing the ϵ-carbide. But at higher tempering temperatures the lower the chromium content the greater is the hardness. This is probably because with increasing chromium content the amount of precipitated V_4C_3 is decreased, with a consequent reduction in the dispersion hardening effect.

Tempering Parameters

From an examination of the tempering characteristics of a wide range of plain carbon and alloy steels, Holloman and Jaffe[26] have suggested that the effects of time and temperature upon the change of hardness on tempering are interrelated. They have proposed an empirical parameter of the form $T(C + \log_{10}t)$, where T is the absolute temperature, C is a constant of value about 20, and t is the tempering time.

In a steel which does not secondary harden, the interdependence of time and temperature may be adequately expressed in the form of this tempering parameter. The results obtained in the case of the

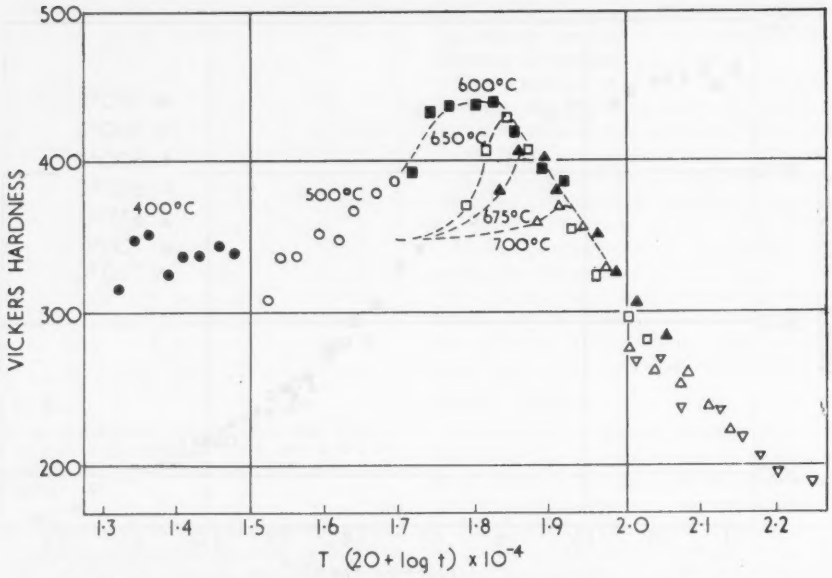


Fig. 24—Change of hardness with tempering parameter for the 0·7% V steel

steel containing 3% Cr are shown in Fig. 23. However, as the value of log t may be small compared with the constant term, the temperature dependence of the parameter is large and, with tempering times of up to 100 h and the temperature intervals used, there is very little overlapping of the hardness values between the respective temperatures ranges.

If secondary hardening is considered as a form of precipitation hardening then the interdependence of time and temperature upon the change of hardness with tempering can never adequately be expressed in the form of the tempering parameter proposed by Holloman and Jaffe. With a precipitation hardening system, the maximum hardness attained on ageing decreases as the temperature is increased. Thus a short time at a high temperature is not equivalent to a longer time at a low temperature. This effect is clearly observed with the 0·7% V steel when the hardness values are plotted against tempering parameter as shown in Fig. 24.

With quenched plain carbon steels where the microstructural type remains the same (carbide particles at the boundaries of ferrite grains) during tempering in the temperature range 400–700° C, it is justifiable to use a tempering parameter to relate time and temperature. But with alloy steels changes in microstructural type may occur on tempering in different temperature ranges which may not be reflected in the corresponding hardness values. An example of this behaviour is found in the 0·7% V steel. This steel could be tempered to the same hardness by a long treatment at 600° C or a short treatment at 700° C. After the former treatment the ferrite grain size would be large as recrystallization would have occurred; in the latter case the ferrite grain size would be small because of the rapid grain boundary precipitation of V_4C_3 at the high temperature. Thus, although the steel may have the same hardness after the two tempering treatments, the properties which are grain-size dependent are quite different.

Microstructure and Creep Properties

The microstructural features which are thought to influence creep properties of metals and alloys have been discussed by Freeman and Corey[27] and it is now generally accepted that finely dispersed precipitates are one of the factors which lead to good creep resistance in an alloy. Correlation between the microstructures of the steels examined in this investigation and the creep data obtained by other workers on similar steels shows that good creep resistance is associated with a finely dispersed precipitate of vanadium carbide.

The creep resistance of the steels on testing at 600° C is improved by a preliminary tempering treatment at 675–700° C. After this heat-treatment the microstructures of the steels containing vanadium show a Widmanstätten precipitation of vanadium carbide within the ferrite grains. Ward and Rait[20] have suggested that the good creep resistance of vanadium steels is a result of coherency at the interface between the carbide and the ferrite matrix. In view of the Widmanstätten type of precipitation the initially formed precipitates of vanadium carbide are undoubtedly coherent with the ferrite matrix, but after tempering at 700° C the precipitate size is such that it is doubtful if the interface is still coherent. It seems more likely that the interface is only partially coherent, but of low energy. Under these conditions the particles will not grow rapidly with an increase in temperature or during the progress of a creep test.

With a steel of base composition 0·2% C–0·5% Mo–0·5% W–0·8% V, Ward and Rait[20] found that the creep resistance was improved further by the addition of 1% Cr but decreased with greater chromium contents to a minimum value at 7%. The electron diffraction patterns from the steels examined

in this investigation, which did not contain tungsten, showed that the vanadium carbide in the absence of chromium precipitated in the form of plates after tempering at 700° C, whilst in the presence of 1% Cr and 1% Mo the carbides were almost spherical. An array of spherical particles in a matrix will hinder the movement of dislocations more than a similar array of plate-like particles. Those plates which are parallel or almost parallel to the planes in which the dislocations move will not provide a very serious obstacle to the passage of the dislocations, whereas spherical particles will be an equally difficult obstacle to a dislocation moving in any plane. Thus chromium, by changing the shape of the precipitated vanadium carbides, may bring about an improvement in the creep properties. A further possibility is that with the addition of only 1% Cr very little Cr_7C_3 is formed on tempering, and therefore a solid-solution hardening effect is to be expected. The decrease of creep resistance with increased chromium content is almost certainly due to the formation of Cr_7C_3 at the expense of V_4C_3 and the consequent decrease in the number of finely dispersed precipitates.

In the 3% Cr steel tempered at 600° and 650° C the number of precipitates of Cr_7C_3 in unit volume of matrix may be as high as that of V_4C_3 in the vanadium steels after tempering at 700° C. However the Cr_7C_3 particles grow rapidly on further tempering, and they may not be as stable as V_4C_3 particles under the conditions of the creep test.

The addition of up to about 1% Mo to a vanadium steel has been shown by Ward and Rait to improve the creep resistance whilst above about 1% Mo the creep resistance is decreased. The micrographs show that no molybdenum carbide is formed with additions of up to 1% Mo unless the vanadium content is low. It is thought, therefore, that the role of molybdenum in improving the creep resistance is largely that of a solid solution hardener. It also probably assists in the spheroidization of the vanadium carbide. When the concentration of molybdenum is high enough to form its own carbide, the amount of vanadium carbide will be reduced, but Mo_2C forms as needles and these may not be as effective as spheres of V_4C_3 in hindering the movement of dislocations.

The influence of grain size upon creep behaviour has been discussed by many workers, but its importance in relation to the creep properties of ferritic alloy steels has not previously been mentioned. From the present investigation it is seen that with vanadium steels and low chromium–vanadium steels it is possible to exert control of the ferrite grain size by controlling the tempering conditions. It may be possible, therefore, to devise suitable heat-treatments for these steels to give improved creep properties, for it appears that when the steels normally go into service the ferrite grain size is small and the grain shape acicular, whereas better creep resistance would be expected from large equiaxed grains.

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A Study of the Tuyere Combustion Zone

By J. Taylor, G. Lonie, and R. Hay

SYNOPSIS

The tuyere combustion zone has been explored on an experimental furnace by means of raceway measurements and gas-composition surveys. The effects of variables such as air velocity and temperature, tuyere diameter, and coke size have been studied. It was concluded that raceway measurements, which are easy to make, are a reliable guide to the extent of the combustion zone.

An expression has been derived for estimating the length of the raceway in terms of the velocity and density of the air, the tuyere diameter, and the coke size. Measurements on operating furnaces were shown to be generally in good agreement with those from the experimental furnace, but the type of tuyere appeared to be an additional factor in raceway penetration.

It was shown that furnace size is the most important factor in raceway penetration so that the active peripheral zone has tended to increase as the diameter of blast-furnaces has increased. 1452

EXPLORATION WORK on the tuyere zone of the blast-furnace has been carried out by a number of workers[1] as part of the wider programme of blast-furnace exploration which followed the pioneer work of Kinney and his colleagues at the U.S. Bureau of Mines. The results for the tuyere combustion zone were rather restricted and confined mainly to gas composition and temperature measurements along the line of the tuyere axis. A controlled study of the variables is difficult on an operational furnace and the only systematic study was that carried out by Russian workers[2] on the effect of blowing rate.

More recently, attention has been directed to the physical movement of the coke particles caused by the jet action of the air blast. The fact that there was a vigorous recirculation of particles in the tuyere zone was revealed by high-speed ciné films taken through the tuyere latch.[3, 4] Systematic studies of the raceway induced in a bed of particles have been made on models by Wagstaff[5] and Paschal and Biver.[6] These studies were carried out under non-combustion conditions and it appeared desirable to study the process under combustion conditions.

EXPERIMENTAL DETAILS

A furnace was constructed to burn coke under conditions similar to those obtaining in the blast-furnace. In designing the furnace, a compromise was sought between one so small that extrapolation of the results up to the scale of the blast-furnace would be dangerous, and one so large that the effort of operating it would be prohibitive. The furnace chosen (Fig. 1) is essentially a rectangular shaft furnace 4 ft wide, 2 ft 6 in. deep, and 8 ft high. Over the top is a hood and chimney to lead away the exit gases. The walls are of high-quality fireclay bricks and the hearth of crushed ganister. The furnace is fitted with three water-cooled tuyeres of 2·19 in. dia., but they could be reduced by inserting sleeves. The tuyeres are spaced 15 in. apart to correspond roughly to the 6-in. tuyere at 5 ft centres common to the modern blast-furnace. The dimensions were based on Wagstaff's observations that the raceway diameter was roughly proportional to tuyere diameter. Experience showed that mutual interference of the raceways with one another need only be expected if all tuyeres are the maximum size. The air supply was brought from the adjacent blast-furnace plant by a 6-in. lagged main. There was a control valve, orifice meter, and pressure and temperature tappings for each tuyere.

Four series of observations were made, namely:

(i) Raceway measurements under non-combustion conditions
(ii) Raceway measurements under combustion conditions
(iii) Gas-analysis surveys of the combustion zone
(iv) Temperature measurements in the combustion zone.

All investigations were carried out on the central tuyere zone, the function of the outer tuyeres being to obviate undesirable wall effects in the central zone. For the raceway measurements it was appreciated that, because of the relatively small diameter of the tuyeres, it would be difficult to insert a probe through the tuyere without interfering with the air flow. Advantage was therefore taken of the rectangular construction and holes made in the back wall through which probes could be inserted. The technique was simply to use a metal rod as a listening probe. As the probe entered the raceway, the rattle of the circulating coke particles against the rod could be heard, and indeed felt, quite plainly. In this way the limit of the raceway could be located with considerable accuracy. For combustion conditions a water-cooled copper tube proved as satisfactory as the metal rod under non-combustion conditions. Measurements were made only along the tuyere axis for non-combustion conditions, but for combustion conditions a three-dimensional survey was attempted by having a series of holes in the back wall and also by using the sidewall sampling hole described below.

Gas samples were drawn through a ½-in. bore water-cooled copper tube. The first attempts at gas sampling were made through the back wall but it

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was found that the tube choked in a matter of seconds owing to bombardment with drops of slag and fine particles of coke. It was necessary, therefore, to insert the tube at right-angles to the tuyere axis. The sampling hole was made in the end wall at tuyere level and fitted with a ball-and-socket joint through which the tube was inserted. In order to locate accurately the end of the sampling tube, a hinged steel plate was fixed so that the line of the hinges was horizontal and passed through the centre of rotation of the ball (Fig. 2). The vertical angle of the plate then fixed that of the sampling tube and the horizontal angle was measured on the plate. These two angles and the length of tube projecting into the furnace located the point at which the sample was taken. The exact settings were checked by trials in the empty furnace.

The problem of temperature measurement is much more difficult than that of raceway measurement or gas sampling and no really satisfactory method has been evolved. Anything in the nature of a suction pyrometer was excluded because of the high dust content of the tuyere-zone gas. A considerable number of readings was taken with a bare thin wire Pt/Pt–Rh couple inserted through the sampling tube but it was concluded that the results must be regarded as suspect. It was impossible to ensure that the junction projected clear of the water-cooled tube because of bombardment by the circulating coke. The results therefore tended to be low. The most reliable results were obtained with a combination of a graphite sighting-tube and optical pyrometer. This worked well for the non-oxidizing conditions of the periphery of the tuyere zone but for the oxidizing conditions nearer the centre the graphite sighting-tube was consumed far too rapidly.

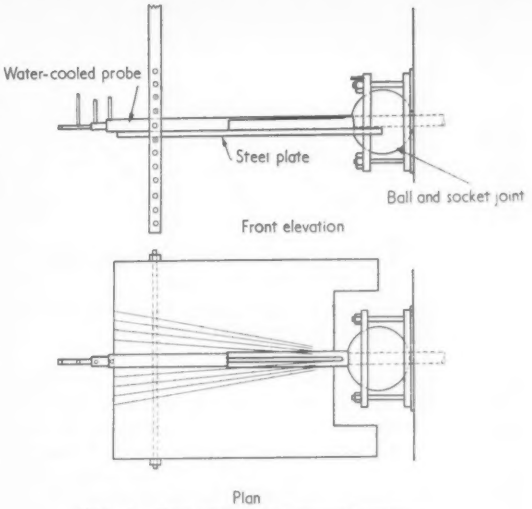

Fig. 2—Gas-sampling arrangements

In operating the furnace, debris from the previous run was first raked out. A fire was started at the rake-out holes and a bed of large coke built up to about the tuyere level. This was left burning under natural draught overnight in order to ensure a thorough heating up of the hearth zone. In the morning the bed was built up with coke of the size to be investigated, the rake-out holes closed, and air turned on at the tuyeres. Steady conditions were very quickly attained and measurements could be commenced within about ½ h of turning on the blast. Preliminary runs had shown a tendency for the high-melting coke ash to form a semi-fused shell round the combustion zone. After cooling, this shell could be recovered and gave a very good picture of the shape of the combustion zone. Such a shell must be expected to impose a restriction on the movement of the coke, and in order to avoid this, fluxes in the form of basic slag and sand were added to give a slag of minimum melting point. With these conditions, a thin fragile shell was obtained only occasionally.

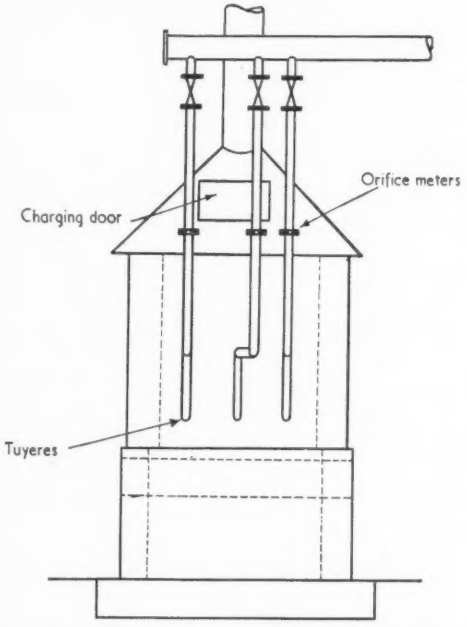

Fig. 1—Front elevation of experimental furnace

RESULTS

Raceway Measurements for Non-combustion Conditions

These were all made along the tuyere axis and represent the parameter measured from the tuyere nose to the limit of penetration. The raceway appears to be less stable for non-combustion than for combustion conditions and the reproducibility of results was not very good. The variables studied were flow rate, coke size, and air temperature at a tuyere diameter of 1 in. In studying the effect of flow rate, the rate was first increased, then decreased, and finally increased again. In most cases there was a fairly pronounced hysteresis effect: i.e. lower values were obtained when increasing the flow rate from zero than when the flow rate was decreasing from the maximum, but those obtained on again increasing

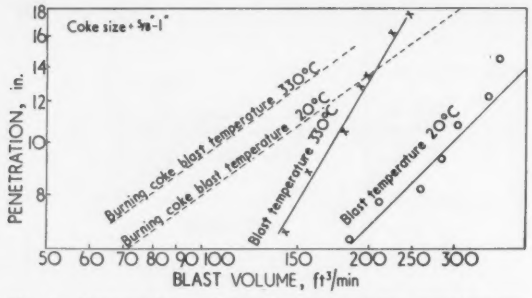

Fig. 3—Comparison of penetrations under combustion and non-combustion conditions

the flow rate coincided fairly well with the second series. It was considered that in the first series the raceway diameters tended to be smaller because the initial resistance of the bed had to be overcome; the recorded values were therefore the mean of the second and third. A further factor to be taken into account was that the self-abrasion of particles revolving in the raceway was very pronounced and there is a sizing effect, smaller particles tending to concentrate at the periphery and larger particles at the centre. There is a notable difference between combustion and non-combustion conditions in this respect. In the former, coke is consumed and new coke is continually feeding into the raceway, whereas in the latter the same particles are circulating the whole time. This is why, for instance, only closely sized particles may be studied under non-combustion conditions. To minimize abrasion and sizing effects, measurements were carried out as expeditiously as possible; but even so it was considered that the results must be influenced to some extent by these factors.

The results obtained were plotted on a log-log scale as penetration/volume-flow (measured at N.T.P.). One such plot, Fig. 3, shows that the values lie fairly well on a straight line. The graphs for all the results are shown in Fig. 4, where for the sake of clarity the experimental points have been omitted. Gradients taken from the curves in Fig. 4 are given in Table I. These show some variation and those for the 1–1¼-in. coke are higher than the others. Wagstaff observed that a raceway could not be initiated if the particle diameter was greater than the tuyere diameter. This was not our experience but we did

Table I

GRADIENTS TAKEN FROM CURVES IN FIG. 4

Coke Size, in.	Slopes of Curves (log ρ/log V)	
	Air Blast Temperature	
	20° C	330° C
1–1¼	2·1	1·8
⅝–1	1·4	1·75
½–⅝	1·6	1·5
¼–½	1·55	1·5

observe that these raceway measurements were much more erratic than for the smaller sizes of coke. Omitting the 1–1¼-in. values, the average gradient is 1·55, i.e. the penetration or raceway diameter is proportional to $V^{1\cdot55}$ whereas Wagstaff found it proportional to V^2. If the linear velocity through the tuyere is used it is found that the results for the three smaller coke sizes and the two air-blast temperatures are fairly well represented by the expression

$$P = B \cdot \frac{\rho}{d} \cdot V_l^{1\cdot55}$$

where P = penetration, in.
ρ = density of air blast relative to air at 20° C and 1 atm
d = mean sieve size of coke particles, in.
V_l = velocity, ft/min.
B = a constant.

If the results are plotted according to this equation there is considerable scatter about the mean line, probably due to the low accuracy of the measurements. It appears, however, that results on cold models are less important than had been thought, since, as shown in the following section, they appear to bear little relation to those for combustion conditions. The difference between non-combustion and combustion conditions is illustrated in Fig. 3. Under combustion conditions much greater penetrations are obtained.

Raceway Measurements for Combustion Conditions

The main series was, as before, the penetration measured from the nose of the tuyere along its axis but, in addition, the shape of the raceway was explored by inserting the probe at different levels in the back and side walls. As the raceway is more stable when coke is being burnt, there is not the same urgency to complete the measurements as in the former series.

The same variables as for the non-combustion series and, in addition, the effect of tuyere diameter and coke size range were studied. Under combustion

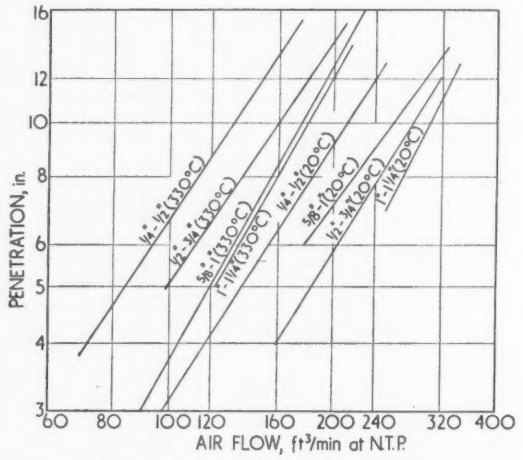

Fig. 4—Air-flow/penetration under non-combustion conditions

conditions there was little or no hysteresis effect except with the largest coke and the 1-in. tuyere, and measurements were much more reliable and reproducible. The results have been plotted as P/V on log-log co-ordinates and those showing the effect of tuyere size are shown in Fig. 5. In all cases the gradients were very close to 0·7 and the relationship can be expressed as

$$P = B'V^{0\cdot7}.$$

The results can then be compared by listing values of B' as in Table II.

A general equation similar to that for the non-combustion results and including the tuyere diameter (D_J) as a variable is

$$\frac{P}{D_J} = \sqrt{\frac{\rho}{d}}\, BV^{0\cdot7}$$

The results for all experiments are shown in Fig. 6. The mean line gives a value of 0·0075 for the constant B. Only the series with coke of wider size range shows any systematic deviation from the mean line, but these results tended to be slightly, but significantly, lower. There are two possible reasons for this:

(i) d has been evaluated by taking the arithmetic mean size whereas the surface or volume means might have been used. When the size range is small the various means are virtually the same, but they diverge as the size range increases. The surface and volume means are larger than the linear mean and, if used, would bring the results for the wider size range more into line with the others.

(ii) There is a pronounced sizing action in the raceway and smaller particles are expelled so that the mean size of the circulating particles is increased. This effect must be more pronounced for a wider size range of particles and could also account for the observed deviation. Since it is impossible to decide which is the more important effect, the linear mean size has been retained because of convenience.

In all the experiments the changeover from raceway to non-raceway conditions was noted. As might be anticipated, measurements of the minimum stable raceway size tended to be rather erratic. They did however show definite trends. For example, increase in this minimum was approximately proportional to the increase in mean coke size. The minimum raceway diameter also increased with tuyere diameter, although much less rapidly, and the ratio P/D_J actually decreased, as did the linear air flow corresponding to the minimum. As noted later, the minimum values of P/D_J are smaller on operating furnaces despite the increase in coke size.

Measurements were made of the maximum height and width of the raceway, the former by inserting the listening probe through holes in the back wall and parallel to the tuyere axis. Although these holes were staggered about the line of the vertical axis of the tuyere, the vertical intervals between holes varied from ½ to 2 in. which did not give sufficient accuracy. Because of furnace construction, the maximum depth below the tuyere axis which could be determined was only 6½ in. and in many cases the raceway extended below this. The most complete series were obtained for ⅜–⅝-in. coke at the two air-blast temperatures (Fig. 7). It is shown that the raceway is approximately symmetrical in this plane and the height increases steadily with blowing rate. For these conditions the total height agrees very well with the penetration. Incomplete results for other coke sizes tended to confirm this. Only the height above the tuyere was measured but, as shown in Table III, this is about half the penetration.

In the horizontal plane the width of the raceway was measured at points 2 in., 8 in., and 12 in. along the axis from the tuyere nose. For any particular

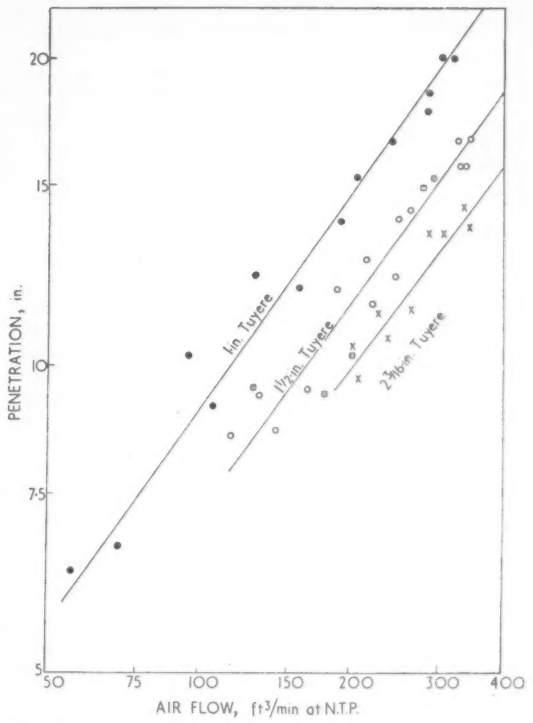

Fig. 5—Air-flow/penetration under combustion conditions

Table II

VALUES OF B' IN EXPRESSION $P = B'V^{0\cdot7}$ (COMBUSTION CONDITIONS)

Coke Size, in.		Blast Temperature, °C	Tuyere Size, in.	B'
Range	Mean			
⅜–⅝	0·5	20	1	0·35
⅜–⅝	0·5	330	1	0·44
⅝–1	0·83	20	1	0·36
⅝–1	0·83	330	1	0·38
1–1¼	1·125	20	1	0·26
⅝–1¼	0·83	20	1	0·31
¼–1	0·625	20	1	0·30
⅝–1	0·83	20	1·5	0·28
⅝–1	0·83	20	2·19	0·24

Table III
BLOWING RATE AND RACEWAY DATA

Coke Size, in.	Flow Rate, ft³/min	Temperature, °C	Raceway Height, in.	Raceway Penetration, in.
⅜–1¼	250	20	6·5	13
1–1¼	300	20	7·0	13·5
1–1¼	150	330	6·5	11

Table IV
HALF-WIDTH VALUES

Coke Size, in.	Tuyere Diameter, in.	Air Blast Temp., °C	Raceway Half-width, n.
⅜–⅝	1	20	2·3
⅜–⅝	1	330	2·7
⅝–1	1	20	2·0
⅝–1	1	330	3·0
1–1¼	1	20	7·0
⅜–1¼	1	20	5·7
⅝–1	1½	20	5·5

set of conditions these half-width values were constant within experimental error and the mean values were taken for the comparison of different conditions. The results were not as consistent as those for linear penetration but variations were random and apparently independent of flow rate. Not only, therefore, is the raceway of uniform width for at least most of its length, but that width is independent of the linear velocity. Half-width values for a number of conditions are listed in Table IV. Although these results are somewhat erratic, they show that coke size and tuyere diameter are the most important factors in determining raceway width.

Measurements on Operating Furnaces

In the derived relationship for estimating the penetration, the effect of blast temperature was included as a factor in the air destiny. On general grounds it was to be expected that the air density should also include the effect of blast pressure. It was impossible to test this on the small furnace, so measurements on operating furnaces were desirable from this point of view and also to test the validity of the relationship under operating conditions. Measurements were made on six different furnaces.

An alternative technique for the measurement of penetration was first required. The only possible method appeared to be the insertion of a probe through the tuyere latch and estimation of the raceway limit by the resistance. Steel and water-soaked

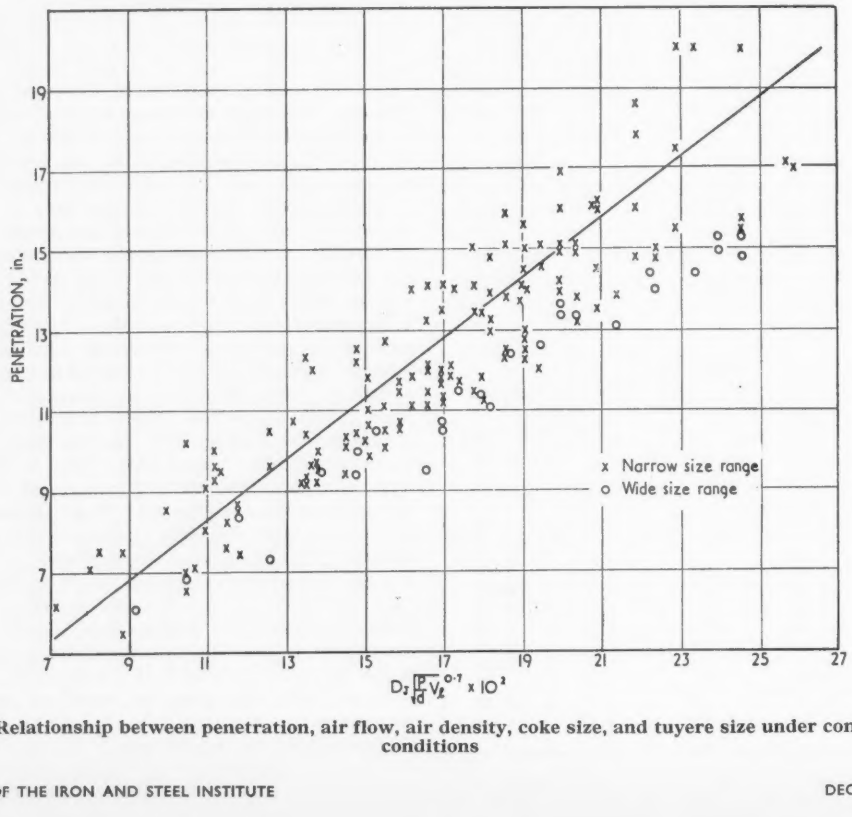

Fig. 6—Relationship between penetration, air flow, air density, coke size, and tuyere size under combustion conditions

Table V

RACEWAY MEASUREMENTS ON OPERATING FURNACES

Furnace	Tuyeres		Air Blast					Mean Coke Size, in.	Tuyere Dia., in.	Calculated Penetration, in.	Observed Penetration		Difference		
	No.	Dia., in.	Volume at N.T.P., ft³/min	Absolute Pressure, lb/in²	Temp., °C	Relative Density	Linear Velocity ft/min				No. of Readings	Mean Value, in.	Difference, in.	% ($\frac{\Delta P}{P}$)	Standard Deviation of ΔP
A1	9 1	6 4·5	38,520	29·4	630	0·65	31,550	2·2	6 dipping	34·4	36	36·9	+2·5	7·3	6·7
A1	9 1	6 4·5	38,830	30·4	580	0·71	29,200	2·2	6 straight	34·1	6	36·0	+1·9	5·6	2·6
A2	9 1	6 4·5	43,870	33·1	600	0·755	30,800	2·2	6 dipping	36·6	67	42·6	+6·0	16·4	10·6
A2	9 1	6 4·5	44,670	35·0	610	0·79	30,000	1·6	6 dipping	43·2	6	48·1	+4·9	11·3	8·8
A2	9 1	6 4·5	46,000	33·7	570	0·80	30,700	2·2	4·5 straight	28·2	4	29·9	+1·7	6·0	4·6
A3	10 2	6 4·5	42,000	29·6	605	0·67	28,600	2·2	6 dipping	32·8	20	34·3	+1·5	4·6	5·2
A3	10 2	6 4·5	42,290	29·7	605	0·675	28,750	2·2	4·5 straight	24·7	12	23·6	−1·1	−4·5	8·9
B2	14 2	6·5 5	73,300	34·6	605	0·79	25,500	2·0	6·5 straight	38·3	20	36·7	−1·6	−4·2	12·3
B2	14 2	6·75 3	74,500	33·0	620	0·74	28,300	2·0	6·75 dipping	40·2	14	40·0	−0·2	−0·5	13·1
B4	12 9	6·5 6·75 (2 at 6·5 blocked)	90,000	37·1	650	0·80	24,800	2·0	6·5 dipping	36·8	37	42·3	+5·5	+15·0	14·3
C3 and C6, common main, each furnace:	12 (2 blocked)	4	32,500	19·7	775	0·37	50,200	2·2	4 straight	24·0	5	26·3	+2·3	+9·6	7·1

wooden poles were tried and the latter in particular proved very satisfactory. Because of the lighter weight and greater ease of handling, the method was more sensitive and determinations could be made more quickly. It was found that the poles entered the combustion zone easily and then came up against what felt like a solid wall; this was taken to be the raceway limit. There is one difference between this method and that adopted on the small furnace. The boundary of the raceway is not a smooth surface, but an irregular wall of coke pieces. In the listening probe method, the probe must intrude clear of this wall before contact is made with the rotating particles. In the resistance method, the pole approaches the wall from inside and can be expected to slip into crevices between the particles. The latter method should, therefore, give rather higher values and the difference should be of the same order as the size of coke, i.e. for 2–3-in. coke it should be ~ 2 in.

The determination of the correct values for blast pressure, blast volume, and coke size raises difficulties on operating furnaces. It is known that there is a drop in pressure between the blower, where the pressure is usually recorded, and the tuyere. According to Russian data[7] this is fairly constant and about 10%. At one of the furnaces where measurements were made the pressure measured in the ring main was 10% lower than at the blower. As the Russian data indicated only a slight drop between ring main and tuyere, a correction of − 10% has been applied to all pressures measured at the blower. Similarly it is known[8] that, owing to losses between the blower and furnace, the metered air volume is always in error by about 10%. However, this is more irregular and no correction has been made to metered volumes. Finally, the coke is subjected to an unknown degree of breakdown between the last point at which the size can be measured and the tuyere zone of the furnace. The modern view is that the breakdown during passage through the furnace is not large and the mean (linear) coke size has been estimated at the furnace bin. Results from operating furnaces are therefore subject to certain systematic errors but, provided these are reasonably constant, comparisons may still be made.

The range of conditions investigated was comparatively restricted as calculations had shown that the range of penetrations to be expected on modern British furnaces lay within a narrow range, viz. 33–40 in. It is true that much smaller values, viz. 24–30 in., could be expected on the older smaller furnaces but frequently in such cases furnaces were blown from a common main. This disadvantage may be surmounted by taking measurements on each furnace in the common main but, even so, data generally are not as reliable as on the more modern furnaces.

The results of a large number of measurements on six different furnaces are summarized in Table V. Because of the varying conditions, the measurements can only be compared by reference to the calculated values. The percentage difference ΔP between the observed and calculated values is therefore shown in the penultimate column, whilst the last column shows the standard deviation of this difference. The recorded standard deviation is rather large in certain cases and that has made more difficult the evaluation of other effects.

It can be shown that there is a significant difference* in the degree of variance of ΔP at the different fur-

* The customary value of 95% probability has been taken as the level of significance.

naces. The calculated variance includes random errors in the data used for the calculated penetration and random errors in the raceway measurement. In view of the general constancy of operating conditions and the fact that the method of measurement was the same, it is considered highly probable that the difference is due to fluctuations in raceway size. Apparently, therefore, raceway conditions are more constant at some furnaces than at others.

It can also be shown that, if the ΔP values for the furnaces with dipping tuyeres are compared, those for A2 (first series) and B4 are significantly greater than those for A1, A3, and B2 (second series). Similarly for straight tuyeres, C3 is significantly greater than B1 (first series). Evidently there is a furnace factor which can influence penetration to the extent of about 10%.

A comparison between dipping and straight tuyeres has also been made. In view of the observation above, this must be made on a furnace by furnace basis. In each case the ΔP value is higher for dipping tuyeres. For the results at furnaces A2 and A3 it is significantly higher, but at A1 and B4 the difference is not statistically significant. This uncertainty is due to the comparatively small number and the large standard deviation of the determinations. Taken together, however, the four pairs of results make it highly probable that dipping tuyeres give a raceway penetration some 5–6% greater than straight tuyeres.

The values of P/D_J on operating furnaces all lie close to the lower limit of the range of results shown in Fig. 6. This is related to the fact that the value of P/D_J at the minimum raceway size decreases with tuyere diameter. For the small experimental tuyeres the lower limit corresponded to about $P/D_J = 7$. On operating furnaces, penetrations as small as 24 in. for a 6-in. tuyere ($P/D_J = 4$) were observed.

DISCUSSION OF RACEWAY MEASUREMENTS

Measurements on operating furnaces covering a fair range of conditions have been shown to be in general agreement with the expression

$$\frac{P}{D_J} = B\sqrt{\frac{\rho}{d}} \cdot V_t^{0\cdot7}$$

derived from the experiments in the experimental furnace. Account must be taken of systematic errors in operating data and the fact that there is a slight difference between straight and dipping tuyeres in evaluating the constant B. It also appears that there is some factor associated with the furnace, and not included in the expression, which influences the penetration to a certain extent but, in general, raceway penetration is described by the above relationship. The effects of various factors on penetration can therefore be deduced from this relationship. These can be more readily appreciated if the expression is re-arranged in the form

$$P = B\frac{V^{0\cdot7}}{\rho^{0\cdot2} \cdot D_J^{0\cdot4} \cdot d^{0\cdot5}}$$

where V is the volume flow in ft³/min at 20° C and 760 mm. Typical variations are illustrated below.

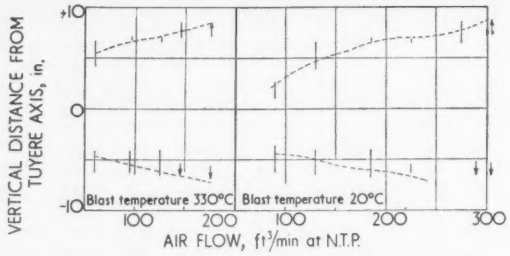

Fig. 7—Maximum raceway height (referred to axis of tuyere) *v.* air-flow for ⅜–⅝-in. coke

Volume of Air per Tuyere—For modern furnaces this may vary from about 3500 ft³/min to about 6500 ft³/min, the latter figure representing an extremely hard-blown furnace, Rice rating about 145. Taking an extreme case of 100% increase this would correspond to approximately 60% increase in penetration. Values for older furnaces were much less, as low as 1000 ft³/min which in many cases corresponded to non-raceway conditions. This therefore represents a notable difference between the older and more modern furnace practice.

Density of Air—This is a function of temperature and pressure. An extreme increase in pressure from, say, 20 lb/in² to 40 lb/in² (absolute) only decreases the penetration by about 11%. A typical range of blast temperature, e.g. 400–850° C, has much the same effect but, of course, the penetration increases with temperature.

Tuyere Diameter—A fairly typical range of tuyere diameter might be taken as 5½–7 in. although smaller tuyeres are frequently used over the tap hole. An increase of 5½–7 in. only corresponds to a decrease in penetration of 11%.

Coke Size—Normally coke mean size does not vary greatly; a typical range of values could be 2·0–2·5 in. at the furnace bin. In the more extreme case of layer charging this might well be increased to 1·5–2·7 in. The former would correspond to a change in penetration of 11% and the latter 34%.

These illustrations show that the penetration is most influenced by the volume flow of air per tuyere which is the dividend of the total air flow by the number of tuyeres. The air blown increases as the cross-sectional area of the hearth,* i.e. as the square of the diameter, whereas the number of tuyeres is limited by the circumference, i.e. varies approximately with the diameter. As furnaces have increased in size, so have the volume of air per tuyere and raceway penetration, a tendency which has been augmented by other changes which have permitted furnaces to be driven harder and also by a trend towards the use of a smaller mean coke size. This trend which has been only slightly modified by the accompanying increase in blast pressure is, therefore, an inevitable feature of increasing furnace size and seems the probable explanation of why earlier predictions as to the limit of hearth diameter have been falsified.

* More exactly as the area of the active periphery of the hearth.

As already noted, normal variations in tuyere diameter have little effect on penetration. The importance of tuyere diameter more probably lies in its influence on the horizontal (and vertical) spread of the raceway. Unfortunately data on this aspect are rather sketchy but there is no doubt that an increase in diameter causes a marked increase in the horizontal width of the raceway. On the assumption that adjacent raceways should almost meet, the tuyere diameter should be controlled by the spacing of the tuyeres and, of course, by the mean coke size. This latter, like the tuyere diameter, has a more pronounced effect on the spread of the raceway than on the penetration. It may be observed that it follows from the above argument that a decrease in coke size should be counteracted by an increase in tuyere diameter and vice versa. Such reciprocal variations would also tend to cancel out the effects on the penetration. Similarly, if the spacing between tuyeres is increased, which would mean a greater air flow per tuyere, the diameter should also be increased so that, as far as penetration is concerned, the two factors tend to cancel each other out.

It is seen that to a great extent the raceway penetration and therefore the active periphery are determined by the hearth diameter. This represents an important difference between the modern furnace and its predecessor. The large furnace shows many advantages, e.g. lower coke rate, more consistent operation, lower sulphur content of iron for a given slag basicity, and, whilst there are certainly other factors involved, the bigger active zone must be taken into account when attempting to ascertain the reasons for the improvement.

The results on operating furnaces showed three interesting features, viz. (*a*) slightly greater penetration with dipping tuyeres as compared with straight, (*b*) a small, but significant difference between furnaces, and (*c*) a difference in the degree of variance of raceway penetration for different furnaces. Dipping tuyeres are employed because they give a longer life. The reason for this is obscure. Straight tuyeres tend to fail on the under side of the orifice owing, one might expect, to the impact of recirculated slag and metal. The effect of dipping tuyeres is to deflect the axis of the raceway downwards, as can be confirmed by viewing the raceway through the latch. Unlike the case of the straight tuyere, no coke can be seen entering the raceway from below. This downward inclination of the raceway might be expected to accentuate attack on the lower lip of the tuyere whereas the reverse is the case. There is likewise no obvious reason why dipping tuyeres should give greater penetration.

In connection with the observation that ΔP may differ slightly for different furnaces, it should be observed that the measurements recorded in Table V refer to normal operation. The period during and immediately after casting was avoided, but otherwise the determinations were randomly distributed and two or more tuyeres were explored. Also, in the case of furnaces A1 and A2 measurements made some months apart agreed very well, thus confirming that the difference was not due to some temporary condition. The differences could, of course, be due to errors in the data on which the calculated penetration was based. This was considered most improbable in the case of furnaces A1, A2, and A3. The same coke was supplied to each furnace and when smaller coke was used on A2 as an experiment, the results were in good agreement with the other values for this furnace. The other possibility was that the wind volume to furnaces A1 and A3 was overestimated by comparison with A2 which would imply that the air loss between blower and furnace was greater. There was no evidence to substantiate this, rather the contrary. Exactly the same argument could be advanced in the case of B2 and B4. The evidence, therefore, supports the view that raceway penetration may be influenced to a slight extent by furnace conditions other than those included in the derived expression.

This view received some confirmation from the observation that, when a furnace was sticking and driving badly, the penetration decreased to as little as $\frac{2}{3}$ of the normal value. Occasionally too, during such abnormal conditions, an exceptionally high value would be obtained. Apparently there was some connection between stock descent and raceway penetration. If so, it seemed possible that there might be a correlation between raceway penetration and operating data during more normal periods. In particular it would be important if raceway measurements could give an early indication of the onset of sticking or hanging. Regular measurements over a period of six days were made to test this hypothesis. The results were inconclusive and no positive correlation could be established between raceway measurements and the occasional periods of poor driving.

The greater variability of the raceway penetration at some furnaces also implies a correlation between stock descent and raceway size. Wagstaff and Holman,[5] who made a similar observation with respect to three furnaces investigated by them, state that the furnace with the smallest variance was also the smoothest working. Smoothness of operation is rather a subjective term and whilst the present data lend some support to the above statement much more extensive data would be required to substantiate this conclusion.

It has been noted that measurements were made on a number of tuyeres and normally the values for different tuyeres were the same. At one furnace, however, there was a period when measurements on tuyeres diametrically opposite to one another were significantly different, viz. 36·9 in. and 30·5 in. These were among the earliest results and the significance could not then be appreciated. The furnace at that time was on rather restricted blast because of labour troubles; later, when conditions became more normal, this difference disappeared.

One way in which conditions in the blast-furnace differ from those in the experimental furnace is the presence of a pool of slag below the tuyeres. Because of insufficient data only a rough estimate can be made of the extent of the raceway below the tuyere, but it is probable that it is approximately half the penetration, i.e. about 18 in. on modern furnaces. The level to which the slag rises relative to the tuyere level is

not easy to estimate. It is a function of the height between slag notch and tuyeres and slagging practice. In the case of the present furnaces the height was $3\frac{1}{4}$-$3\frac{1}{2}$ ft and a slag build-up above the notch to the extent of one ladle could raise the level about 1 ft. That should still leave a small margin between the raceway and slag level and certainly no indication was obtained from the measurements of any interference between slag and raceway. But the fact that the raceway and, therefore, the oxidizing zone can extend so near to the probable slag level is a consideration which should be taken into account when deciding the height between slag notch and tuyere. It should also be borne in mind that the larger the furnace the greater tends to be the penetration and depth of the raceway.

Gas-analysis Survey of Combustion Zone

Exploration of the zone had to be made in three dimensions. In the first four experiments, samples were taken at frequent intervals over the whole of the vertical plane of the tuyere axis and a half of each of three horizontal planes, through the tuyere axis, $4\frac{1}{2}$ in. above, and 4 in. below the axis. It was assumed that the horizontal planes were symmetrical about the axis and only one half needed to be explored. Even so, this exploration involved taking 70–80 samples spread over 5–6 runs, carried out at different times. The results were consistent which showed that raceway conditions were stable and reproducible. Analysis of these results showed that virtually as much information could be derived from the survey of the vertical and horizontal planes through the tuyere axis. In later runs only these were explored.

Results

The method of expressing the results presents some difficulty because there are three important constituents of the gas and the combustion zone is three-dimensional. One way of describing the gas composition is in terms of the oxygen available for oxidizing carbon, etc.; calling this O_g, then $O_g = (2 \times \%O_2) + \%CO_2$. If values for O_g are plotted against $O_2\%$ and $CO\%$, as in Fig. 8, it can be seen that there is a slight overlap between the decreasing oxygen curve and the initial appearance of CO. All results lie very close to the mean curve so that from quoted values of O_g the corresponding O_2, CO_2, and CO percentages can be obtained from Fig. 8 within experimental error.

Values of O_g were first plotted for each plane. Convenient iso-O_g lines, e.g. 18, 10, and 5, were then drawn; the 18 contour represents the approximate limit of penetration of free oxygen and 5 is the smallest value which could be estimated with reasonable accuracy. From these contours the shape of the combustion zone may be envisaged. The effects of the variables studied are illustrated in Figs. 9 and 10.

Figures 9*a*, *b*, and *c* show the effect of increased blowing rate. Figure 9*a* represents non-raceway conditions, the flow rate (96 ft³/min) being just below the critical flow ($\sim$ 100 ft³/min) at which the raceway begins. There is no marked change between non-raceway and raceway conditions, but rather a progressive change with flow rate. The general effect of increased air flow is to elongate the combustion zone, without much increase in either height or width, and displace the maximum away from the tuyere zone. The same effect is shown by the series for air blast preheated to 330° C. In this series and for one experiment with cold air a Durham coke was used instead of the Scottish used in the rest of the experiments. The result for cold blast was practically identical with the corresponding one for the Scottish coke. A direct comparison is therefore possible between the hot blast results and the main series. Reference to Table VI shows that the volume of the oxidizing gas zone is much reduced when using preheated air.

The effect of coke size and range is illustrated by Fig. 10. A comparison of Figs. 9 and 10*a* shows that an increase in size from $\frac{5}{8}$–1 in. to 1–$1\frac{1}{4}$ in. results in a considerable broadening and heightening of the combustion zone and the maximum cross section has moved back to the tuyere nose as in the lower flow rates of Fig. 9. This latter effect is accentuated for a wider size range $\frac{3}{8}$–$1\frac{1}{4}$ in. (Fig. 10*b*).

The effect of an increase in tuyere size is shown by Figs. 9*c* and 10*c*. It is very similar to that of an increase in particle size, i.e. the height and width of the zone are much increased and the maximum is brought back to the wall of the furnace.

A more quantitative comparison of the dimensions of the combustion zones is shown in Table VI. There the areas enclosed by the 18, 10, and 5 O_g contours in both the horizontal and vertical sections are shown. Relative values of the volumes enclosed by the 18 and 10 contour lines are also shown. These are calculated from the relationship

$$\text{Volume} \propto (\text{mean cross-sectional area})^{3/2}$$

on the assumption that the shape factor is constant. This is approximately true for the conditions studied. Lastly the linear penetration as measured by the 18, 10, and 5 O_g contours is compared with the linear penetration from the raceway measurements.

The various attempts to carry out temperature surveys to supplement the gas-composition surveys were not very satisfactory. Measurements could not be made within the oxidizing zone of the raceway because of the rapid wastage of the carbon sighting tube. A comparison between the two air-blast temperatures 20° C and 330° C and the same raceway size showed a difference of about 170° C (1730 — 1560° C) just within the raceway and about 90° C (1570 — 1480° C) 2–3 in. outside the raceway limit. The calculated increase in combustion temperature is about 200° C. Apparently, therefore, an increase in blast temperature raises the temperature within the raceway by about the calculated amount. The rapid decrease outside the raceway may be ascribed to cooling effects in the comparatively small furnace.

DISCUSSION

These results show that for cold air the penetration as measured by the 5 O_g contour is on average 10% greater than that measured by the raceway. Experiment *C* is exceptional in that it is about 10% lower, but in this case the flow rate and penetration were

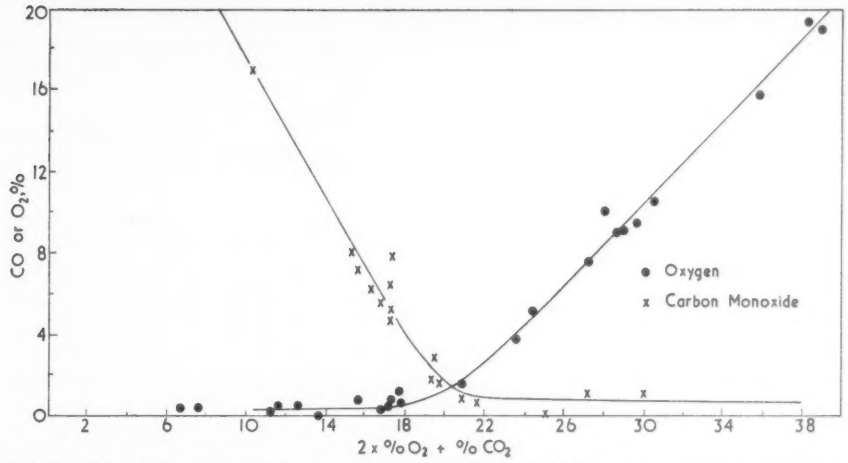

Fig. 8—Relationship: oxygen and carbon monoxide content of combustion gas *v.* O_g ($2 \times \%O_2 + \%CO_2$)

exceptionally high for the coke and tuyere size. The other variables studied, i.e. make of coke, coke size, and tuyere diameter, may have some effect but, if so, it is not very pronounced. Blast temperature does, however, have a very definite effect as shown by experiments *J*, *K*, and *L*, where the 5 O_g penetration

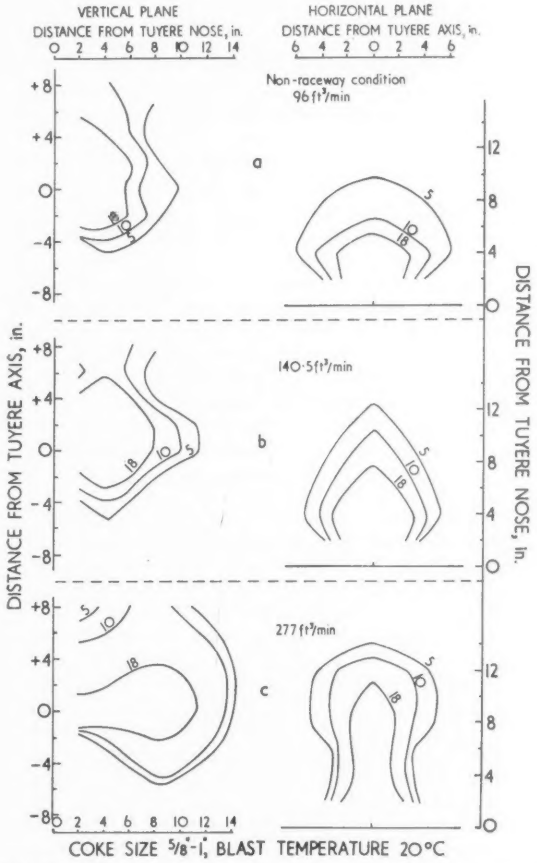

Fig. 9—Effect of air flow on shape of combustion zone shown by iso-O_g contours

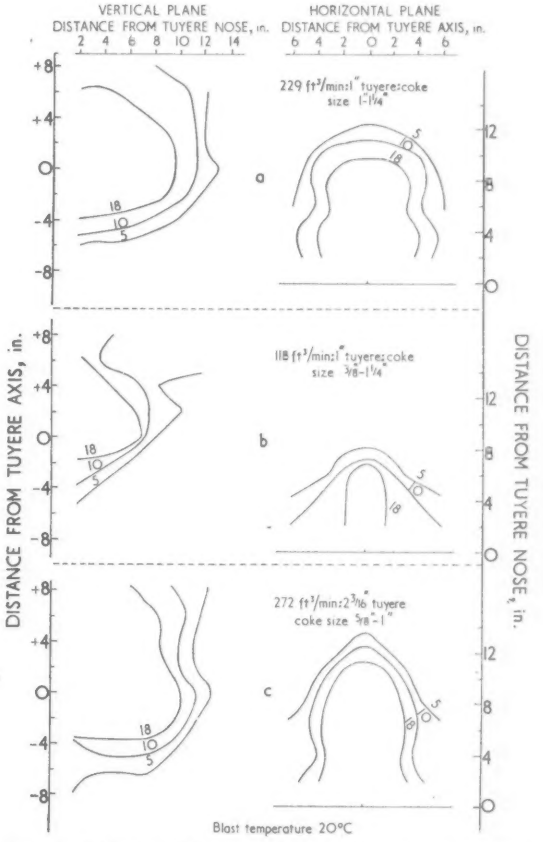
VERTICAL PLANE
DISTANCE FROM TUYERE NOSE, in.
HORIZONTAL PLANE
DISTANCE FROM TUYERE AXIS, in.
229 ft³/min:1" tuyere:coke size 1-1¼"
118 ft³/min:1" tuyere:coke size ⅜"-1¼"
272 ft³/min:2 3/16" tuyere coke size ⅝"-1"
DISTANCE FROM TUYERE AXIS, in.
DISTANCE FROM TUYERE NOSE, in.
Blast temperature 20°C

Fig. 10—Effect of coke size on shape of combustion zone shown by iso-O_g contours

Table VI

COMPARISON OF DIMENSIONS OF THE COMBUSTION ZONES AND OF LINEAR PENETRATION AS MEASURED BY THE O_g CONTOURS WITH VALUES DERIVED FROM RACEWAY MEASUREMENTS

	Coke Size, in.	Flow Rate at N.T.P. ft³/min	Flow Rate at N.T.P. ft/min	Air Temp., °C	Tuyere Dia., in.	Areas enclosed by O_g contours, in² — Vertical Plane 18	10	5	Horizontal Plane 18	10	5	Vol. enclosed by O_g contours 18	10	Linear Penetration of O_g contours, in. 18	10	5	Linear Penetration of Raceway, in.
Clyde coke	⅝–1	118	21,600	20	1	40	70	...	21	45	66	170	440	7·0	8·0	11·0	9·0
	,, ,,	140·5	27,500	,,	1	40	100	...	35	55	70	230	680	7·5	9·7	11·3	10·6
	,, ,,	277	50,800	,,	1	45	125	...	40	80	110	280	1030	11·0	13·0	14·0	16·4
	1–1½	229	42,000	,,	1	65	140	...	65	100	130	580	1320	9·5	11·2	12·2	12·1
	⅝–1¼	118	21,600	,,	1	45	70	...	20	60	...	180	520	7·0	7·7	8·7	8·5
	⅝–1	135	11,000	,,	1·5	74	125	...	52	80	100	500	1100	7·2	9·2	10·8	8·4
	⅝–1	272	10,400	,,	2·19	>110	...	...	62	82	...	...	...	10·5	11·5	12·5	11·8
	⅝–1	96	17,600	,,	1	30	70	...	25	40	92	145	510	5·5	7·0	9·5	Non-raceway
Consett coke	⅝–1	118	21,600	20	1	45	80	...	35	55	70	250	540	7·2	9·5	10·5	9·5
	,, ,,	81·5	30,700	330	1	20	45	70	22	35	50	90	250	4·5	5·5	6·0	8·4
	,, ,,	123	46,000	330	1	35	45	70	20	35	55	100	250	5·7	7·0	8·8	11·3
	,, ,,	182	68,500	330	1	30	65	115	33	60	80	180	490	8·0	10·5	12·0	14·3

is about 20% less than the raceway penetration. This means that for the same raceway size the oxidizing zone is relatively smaller with hot blast than with cold. The significance of this is not easy to assess. Since mixing of air and coke should be similar in the same size of raceway, it would imply that the specific rate of reaction between coke and oxidizing gas is greater at the higher temperatures which are attained with preheated air. This is contrary to the usual view that under these conditions specific reaction rates are so high that the contact between gas molecules and carbon is the rate-controlling factor. As far as these results are concerned the difference seems to be that with cold air combustion of coke continues for a short distance beyond the limits of the raceway, but with preheated air combustion is probably* completed within the raceway. In the latter case it is immaterial whether oxidizing gas has been used up well within the limits of the raceway or not; all the circulating coke is burning so that the raceway effectively delineates the zone in which combustion of coke takes place.

The gas-composition results confirm the raceway measurements as to the effect of various factors on the width of the tuyere zone. This increases with coke size and tuyere diameter but is hardly affected by flow rate. As for the raceway measurements, the height of the tuyere zone is also shown to increase with coke size and tuyere diameter. The effect of flow rate on the height of the zone is, however, more indefinite. Figure 9 shows that the vertical plane of the combustion zone is asymmetric at low flow rates owing to the tendency of the oxidizing gases to extend upwards. Increase in flow rate gives a more symmetrical zone so that although the depth below the tuyere increases slightly as expected from raceway measurements, this is more than offset by a decrease in height above the tuyere.

* The limit of combustion should be when $O_g = 0$, whereas for practical reasons the 5 O_g contour has been mapped.

CONCLUSION

The general conclusion to be drawn from this investigation is that in the large modern blast-furnace the tuyere combustion zone is characterized by a raceway and that this raceway effectively demarcates the zone in which consumption of coke and reoxidation of iron, silicon, etc. takes place. In this respect the modern furnace differs from its smaller predecessor which frequently did not develop a raceway. The effect of coke size in the two cases must be quite opposite; in a stationary bed the penetration of oxidizing gas decreases with decrease in coke size whereas under raceway conditions penetration increases. This offers an argument in favour of the use of smaller sized coke in modern furnaces.

The correlation between raceway and combustion zone is important because the raceway can be investigated comparatively easily and the results are directly applicable to the combustion zone. An interesting deduction from the experimental data is that the penetration and depth of the raceway tend to increase with furnace size and operators might consider whether there is sufficient height between slag notch and tuyere belt in the larger furnaces. There is a strong suggestion that the small but significant differences between furnaces, both as regards the magnitude and variability of the penetration, are associated in some way with stock descent. Insufficient data were obtained to establish a definite correlation, but penetration measurements are so simple that systematic measurements over a period of time might well yield useful information.

Acknowledgments

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An Investigation of a Method for the Combined Determination of Niobium and Tantalum in Steel

By the B.I.S.R.A. Methods of Analysis Committee

SYNOPSIS

A collaborative investigation has been carried out of a method for the combined determination of niobium and tantalum in steel. 1474

IN 1953 THE HIGHLY ALLOYED STEELS and Ferro-alloys Analysis Sub-committee was asked to undertake the development of a method of analysis for niobium and tantalum in alloy steels. A Study Group was formed to formulate a method for examination by the Sub-committee and it was recommended that the problem of the separate determination of niobium and tantalum should be deferred until a method for the combined determination of the two elements in steel had been examined and approved. Similarly analysis of ferro-alloys was deferred. Work on the combined determination of the two elements had already been published[1, 2] and it was agreed that this should be used as the basis of work by the Study Group.

EXPERIMENTAL DETAILS

An experimental trial of a method published by Bagshawe and Elwell[1] was undertaken by the Study Group using four alloy steels, the approximate compositions of which are given in Table I.

The procedure provided for the presence of tungsten, titanium, and molybdenum; by a slight adjustment of conditions, the tungsten precipitate could be made quantitative and a simultaneous determination made. The principles adopted in the method included hydrolytic separation from a solution of low acid concentration in which iron is reduced to the divalent condition with an excess of sulphurous acid, and the precipitate resulting from this hydrolysis may be used to give simultaneous recovery of tungsten by treatment of the solution with aqueous Rhodamine B. The mixed residue is then treated by an adaptation of the

Table I

APPROXIMATE PERCENTAGE COMPOSITION OF THE STEELS USED

Sample No.	C	Mn	Si	Ni	Cr	Mo	Co	W	Cu	Al
234	0·1	0·5	0·5	8·5	1·8	...	...	...	...	...
235	...	...	...	8·0	18·0	3·0	...	...	...	...
236	...	1·0	1·0	13·5	13·5	1·5	10·0	2·5	...	...
237	...	...	...	...	18·5	...	12·0	...	3·23	9·0

Paper MG/D/129/57 of the Methods of Analysis Committee of the Metallurgy (General) Division of the British Iron and Steel Research Association, received 13th August, 1957.

The constitution of the Highly Alloyed Steels and Ferro-alloys Analysis Sub-committee at 31st March, 1957, was: Mr. B. Bagshawe (Chairman), Brown-Firth Research Laboratories; Mr. C. D. Atkinson, Sheffield Testing Works Ltd.; Mr. J. W. Flint, William Jessop and Sons Ltd.; Mr. L. E. Gardner, Edgar Allen and Co. Ltd.; Mr. S. Harrison, Kayser Ellison and Co. Ltd.; Mr. J. D. Hill, Bragg Laboratory, N.O.I.D.; Mr. G. M. Holmes, London and Scandinavian Metallurgical Co. Ltd.; Mr. G. Murray, Brown Bayléy Steels Ltd.; Mr. R. T. Postlethwaite, Samuel Fox and Co. Ltd.; Mr. L. N. Taylor, Samuel Osborn and Co. Ltd.; Mr. J. L. West, Hadfields Ltd.; and Mr. W. Wright, Permanent Magnet Association.

The Sub-committee acknowledges with thanks assistance in this project received from Mr. C. L. Grayson, English Steel Corporation Ltd. at the time of the investigation, now with Edgar Allen and Co. Ltd.

magnesia separation[3] following fusion with potassium carbonate. This gives a quantitative yield of niobium and tantalum, together with effective separation from tungsten and molybdenum. There is no complete separation from titanium, this element being determined colorimetrically in the final residue.

It was suggested that the method should be amended by the addition of 10 ml perchloric acid to the initial solution for samples 235 and 236 and the results obtained after making this addition are reported. It was found, however, that perchloric acid was not necessary and the addition was omitted from the final procedure.

The results obtained by the Study Group, using the method put forward by Bagshawe and Elwell, were thought to be quite satisfactory (*see* Table II) and a further investigation was then carried out by the whole Sub-committee. The results of this examination are reported in Table III.

On the basis of this experimental work, the method was approved for the combined determination of niobium and tantalum in steel and the attention of the Study Group was then turned to the problem of the separate determination of these two elements. This programme of work is still being carried out and the preliminary results were reported in a paper given at the 11th Chemists' Conference in October, 1957.

METHOD

Special Solutions

Magnesium sulphate precipitating agent (*A*)—10 g of magnesium sulphate and 20 g of ammonium chloride were dissolved in about 150 ml of water. After cooling, 4 ml of ammonium hydroxide (50% v/v) were added, and the mixture diluted to 250 ml.

Wash liquor (*B*)—10 g of magnesium sulphate and 10 g of ammonium chloride were dissolved in about 150 ml of water. After diluting to nearly a litre with warm water, 10 g of anhydrous potassium carbonate were added, and the mixture shaken until it dissolved. 8 ml of ammonium hydroxide (50% v/v) were then added and the flask shaken again until the solution was clear. Adjustment to 1 l. was made.

The quality of all reagents should be the best obtainable.

Procedure

For steel containing less than 2% Nb a 5-g sample was weighed. For contents exceeding this amount 2-g samples were used. The sample was transferred to a 600-ml beaker and 60 ml of hydrochloric acid (sp.gr. 1·16) was added. When the metal portion had dissolved, nitric acid (sp.gr. 1·42) was added, drop by drop, to complete oxidation and a slight excess. The solution was then evaporated to dryness and baked lightly for approximately 5 min (*Note* 1). The dry residue was extracted with 30 ml of hydrochloric acid (sp.gr. 1·16) and evaporated to 15–20 ml, care being taken to avoid localized drying out of iron salts, etc. After cooling slightly, 100 ml of sulphurous acid (saturated aqueous sulphur dioxide solution) was added; the solution was mixed by swirling, and heated (*Note* 2). After boiling for 2–3 min to ensure efficient reduction of ferric iron, the solution was diluted to approximately 250 ml with water. A small pad of macerated paper pulp was added and the temperature again raised to boiling and held at this point for 2–3 min. The source of heat was removed and the mixture allowed to settle for at least 30 min (*Note* 3).

It was then filtered through a paper-pulp pad and washed with warm hydrochloric acid (5% v/v). The filter paper and precipitate were transferred to a platinum crucible and ignited cautiously, until organic matter had been removed, then finally at approximately 800° C. After cooling, a few drops of sulphuric acid (50% v/v) and approximately 2 ml of hydrofluoric acid (40% w/v) were added and it was evaporated to dryness, preferably under a radiant heater, and re-ignited at approximately 800° C for a further 2–3 min (*Note* 4) and cooled. The residue was fused with 4 g of potassium carbonate, the cold melt extracted with warm water, boiled for a few minutes, and then the alkaline extract filtered through a paper-pulp pad, the latter being washed with hot water containing a little potassium carbonate (*Note* 5), heated to 70–80° C, and 25 ml of the $MgSO_4$ precipitating agent (*A*) added. After an hour on a water bath, the niobium precipitate was filtered through a paper-pulp pad and washed with warm wash liquor (*B*).

The pad and precipitate were transferred to a beaker, 40 ml of hydrochloric acid (50% v/v) added, and 10 min allowed to extract magnesium salts, etc. After diluting to approximately 200 ml with water, 25 ml of 20% ammonium chloride solution and 20 ml of sulphurous acid were added and the solution was heated almost to boiling. It was next cooled slightly, made just ammoniacal, transferred to a hot water bath for 15 min, and then filtered through a paper-pulp pad, washing with hot 10% ammonium chloride solution. The paper and precipitate were ignited in a platinum dish and weighed as a mixture of $(NbTa)_2O_5$ and TiO_2 (*Note* 6). The residue was fused with 3 g of fused potassium bisulphate and extracted with approximately 75 ml of water containing 7·5 g of tartaric acid. 10 ml of hydrogen peroxide (20 vol) was added and titania determined colorimetrically. The weight of titania (TiO_2) was deducted from the

Table II

RESULTS OBTAINED BY THE STUDY GROUP, Nb(+ Ta)%

Analyst	Sample 234	Sample 235	+$HClO_4$	Sample 236	+$HClO_4$	Sample 237
1	0·30	1·06	...	3·10 3·12 3·19		1·19 1·21
2	0·29 0·27 0·29	1·02 1·00 1·00	0·98	2·96 3·00 2·98	2·98	1·14 1·14 1·13
3	0·28 0·28	0·99 1·01	1·01 1·01	3·01 3·00		1·06 1·08
4	0·275	1·02	1·02	3·05 2·98	2·96 2·94	1·12
5	0·30 0·28 0·28	1·00 1·03 ...	1·02	3·07 3·05 ...	3·03	1·18 1·20 1·20
6	0·27 0·27 0·27	1·00 1·01 ...	1·00 1·00 ...	3·02 3·04 ...	3·00 3·00 ...	1·12 1·14 1·14

Table III

RESULTS OBTAINED BY THE SUB-COMMITTEE, Nb(+Ta)%

Analyst	Sample 234	+$HClO_4$	Sample 235	+$HClO_4$	Sample 236	+$HClO_4$	Sample 237	+$HClO_4$
1	0·28	0·27	1·06	1·09	3·03	3·16*	1·14	1·13*
	0·29		1·06		3·05	3·14	1·13	
	0·30	0·28	1·06				1·10	
2	0·29	...	1·02	0·98	2·96	2·98	1·14	...
	0·27		1·00		3·00		1·14	
	0·29		1·00		2·98		1·13	
3	0·28	...	0·99	1·01	3·01	3·02	1·13	...
	0·28		1·01	1·01	3·00	3·02	1·12	
4	0·275	...	1·02	1·02	3·05	2·96	1·12	...
					2·98	2·94		
5	0·30	...	1·00	1·02	3·07	3·03	1·14	...
	0·28		1·03	...	3·05	...	1·15	
	0·28		...	...	...	...	1·16	
6	0·27	...	1·00	1·00	3·02	3·00	1·12	...
	0·27		1·01	1·00	3·04	3·00	1·14	
	0·27		...	...	...	...	1·14	
7	0·32	...	0·97	1·02	2·85	2·90	1·12	...
8	0·25	...	1·00	...	2·96	...	1·12	...
	0·25		0·97	(chromium removed)		...	...	...
9	0·29	...	1·01	...	3·01	...	1·16	...
10	0·29	...	1·07	...	3·04	...	1·07	...
	0·28		1·02		3·11		1·13	
Average	0·28		1·02		3·01		1·13	

* Examination of the $HClO_4$ precipitates by the A.S.T.M. method revealed that the recommended single treatment for removal of WO_3 and MoO_3 does not appear to effect a complete removal and the figures now given are after a double purification treatment

combined weight of $(NbTa)_2O_5 + TiO_2$ and the difference calculated in terms of niobium or tantalum using the appropriate factor (*Note* 7):

$$Nb_2O_5 \times 0{\cdot}6990 = Nb$$
$$Ta_2O_5 \times 0{\cdot}8189 = Ta$$

Notes

(1) High-temperature baking should be avoided as this increases the tendency to form basic chromium compounds.

(2) The use of a large excess of sulphurous acid ensures efficient reduction of ferric iron and assists the complete hydrolysis of niobium.

(3) A longer settling period does no harm.

(4) Silicon may be determined at this stage from the loss in weight resulting after hydrofluoric acid treatment. Complete expulsion of sulphuric anhydride must be achieved by ignition to constant weight.

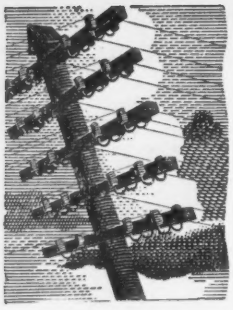
(5) Small amounts of iron precipitated in the potassium carbonate fusion extract do not readily coagulate, and may pass into the filtrate with the niobium. Where a more precise result is required the extract should be treated with a little macerated paper pulp digested on a hot water bath for at least an hour prior to filtration and a repeat fusion of the residue made. None of these refinements is necessary for normal industrial purposes.

(6) The mixed oxide residue does not always contain the total titanium content of the steel. In the presence of niobium, the hydrolysis of titanium is partially inhibited in the initial separation. With very low-titanium steels (less than 0·10%) the total titanium may often be found with the niobium, as much of it is insoluble in the initial acid attack.

(7) If niobium is the predominant element, the niobium factor should be used; the tantalum factor if tantalum is predominant. It should, however, be recognized that some error will arise in using either factor depending on the niobium/tantalum content of the ferro-alloy used in manufacture. If this is known, a more accurate result can be obtained by using a factor calculated from the Nb/Ta ratio of the ferro-alloy.

References

1. B. Bagshawe and W. T. Elwell: *Journal of the Society of Chemical Industry*, 1947, vol. 66, pp. 398–400.
2. S. W. Craven: *Ibid.*, 1947, vol. 66, pp. 400–402.
3. A. R. Powell, W. R. Schoeller, and C. Jahn: *Analyst*, 1935, vol. 60, pp. 506–514.



N E W S

Announcements and News of Science and Industry

THE IRON AND STEEL INSTITUTE

Special Meeting in Belgium and Luxembourg, 1958

The Institute is holding a Special Meeting in Belgium and Luxembourg from **18th to 28th June, 1958.** It will form part of the Journées Internationales de Sidérurgie, organized by the Centre National de Recherches Métallurgiques (C.N.R.M.), the Groupement des Industries Sidérurgiques Luxembourgeoises (G.I.S.L.), and the Groupement des Hauts Fourneaux et Aciéries Belges. H.M. King Baudouin and H.R.H. the Grand Duchess Charlotte have graciously consented to be Patrons of the meeting, which will be under the Honorary Presidency of Monsieur F. Perot (President of the C.N.R.M.), Monsieur L. Bouvier (President of the G.I.S.L.), Monsieur P. van der Rest (President of the Groupement des Hauts Fourneaux et Aciéries Belges), Mr. A. H. Ingen-Housz (President of the Institute), and Mr. C. R. Wheeler, C.B.E. (President-Elect of the Institute).

The general theme of the meeting, which will be divided into three sections, will be "*New Manufacturing Processes for Use in Iron and Steelmaking.*" Section I of the meeting will be held in Liège, from 18th to 21st June, Section II in Luxembourg, from 21st to 25th June, and Section III in Charleroi and Brussels, from 26th to 28th June. The outline programme (subject to alteration) is as follows:

SECTION I—LIÈGE

Theme: New Steelmaking Processes

Wednesday, 18th June
- Opening Session at the Palais des Congrès
- 1st Technical Session—*The Bottom-blown Converter*
- 2nd Technical Session—*The Oxygen-blown Converter*
- Dinner at Casino de Spa, by invitation of C.N.R.M.

Tuesday, 19th June
- 3rd Technical Session—*Steelmaking in the Open-hearth Furnace*
- 4th Technical Session—*Large Electric Arc Furnaces*
- Works visits to: S.A. Cockerill-Ougrée
  C.N.R.M. Laboratories
  Usines à Tubes de la Meuse
- Evening reception and ball by invitation of the Burgomaster of Liège.

Friday, 20th June
- Works visits to: S.A. Métallurgique d'Espérance-Longdoz
  S.A. Cockerill-Ougrée S.A. Phénix-Works
- Evening performance at the Conservatoire Royal de Liège

Saturday, 21st June
- Works visits to: S.A. Métallurgique d'Espérance-Longdoz
  S.A. Cockerill-Ougrée Usines à Tubes de la Meuse

SECTION II—LUXEMBOURG

Theme: New Ironmaking Processes

Sunday, 22nd June
- Free day

Monday, 23rd June
- Opening session at Nouveau Palais des Expositions
- 5th Technical Session—*Low-shaft Furnaces*
- 6th Technical Session—*New Direct Processes*

Tuesday, 24th June
- 7th Technical Session—*Special Blast-furnace Techniques*
- Works visits to: ARBED (Terres Rouges)
  ARBED (Belval)
  ARBED (Dudelange)
- 8th Technical Session—*Blast-furnace Burden Preparation*
- Works visits to: ARBED (Esch)
  HADIR (Differdange)
- Reception and dinner by invitation of G.I.S.L.

Wednesday, 25th June
- Works visits to: ARBED (Terres Rouges)
  ARBED (Belval)
  ARBED (Dudelange)
  HADIR (Differdange)
  Minière et Métallurgique de Rodange
- Reception by invitation of the High Authority of the European Coal and Steel Community.

SECTION III—CHARLEROI AND BRUSSELS

A. Charleroi

Theme: Continuous Casting of Steel

Tuesday, 26th June
- Works visits to: S.A. des Laminoirs, Haut Fourneaux, Forges, Fonderies et Usines de la Providence
  Aciéries et Minières de la Sambre S.A. (A.M.S.)
  Hauts Fourneaux, Forges et Aciéries de Thy-le-Château et Marcinelle S.A.
  C.N.R.M. Laboratories
  Forges de Clabecq S.A.
  C.N.R.M. continuous casting plant at A.M.S.

Friday, 27th June
- Opening session at Palais des Expositions
- 9th Technical Session—*Continuous Casting of Steel*
- 10th Technical Session—*Industrial Application of Continuous Casting*
- Dinner at Palais des Expositions by invitation of C.N.R.M. (Charleroi Section)

B. Brussels

Thursday, 26th June / **Friday, 27th June** } Visits to Brussels International Exhibition

Saturday, 28th June
- Closing session
- Farewell lunch by invitation of the Groupement des Hauts Fourneaux et Aciéries Belges

Full details of the programme, and also of the programme for ladies, will be published in the *Journal* next year, and application forms will be circulated to Members.

NEWS OF MEMBERS

The **President** (Mr. A. H. Ingen-Housz) has been elected an Honorary Member of the South African Institute of Mining and Metallurgy for the year ending 30th June, 1958.

Mr. **N. P. Bacon** has joined Guest Keen Iron and Steel Co., Ltd., as Research Engineer.

Dr. **E. C. Bain** has retired from the United States Steel Corporation.

The Hon. **R. A. Balfour** has succeeded his father as Lord Riverdale of Sheffield.

Mr. **B. C. Bruce Gardner** has been appointed a director of the Patent Shaft and Axletree Co., Ltd.

Mr. **Lewis Chapman,** C.B.E., has been appointed President-Elect of the British Iron and Steel Federation in succession to the late Mr. Gerald Steel, C.B.E.

Dr. **T. E. Dancy** is now Research Supervisor in the Research Laboratory of the Jones and Laughlin Steel Corporation, Pittsburgh.

Mr. **W. C. Fletcher** is joining the Rhodesian Iron and Steel Company Ltd. (RISCO) as General Manager; he takes up his new post on 1st January.

Mr. **J. Hacon** has left the United Steel Companies Ltd. to join the Central Electricity Authority's Leatherhead laboratories.

Mr. **J. H. Hensler** has been awarded the degree of Master of Engineering Science by the University of Melbourne, and is now a Research Officer with Rylands Brothers (Australia) Pty. Ltd.

Mr. **H. P. Hughes,** formerly Works Manager and Technical Adviser to Union Carbide Ltd., Alloys Division, is now Works Manager with Gemec Ltd.

Mr. **F. P. Jansen** is now a chemist with the Messina (Transvaal) Development Company.

Mr. **N. L. G. Lingwood** has been re-elected President of the British Acetylene Association.

Sir **Andrew McCance,** F.R.S., has been invited to continue in office as President of the British Iron and Steel Federation for 1958.

Mr. **S. M. Naim,** now Metallurgical Planning Officer of Hindusthan Steel Private Ltd. (Bhilai Project Section), is at present on a visit to the U.S.S.R.

Mr. **D. J. F. Nancarrow** has obtained an Honours B.Sc. degree in metallurgy at University College, Cardiff, and is now a research student there.

Mr. **J. Savas** is now with Republic Steel Corporation, Canton, Ohio, U.S.A.

Mr. **A. Speirs** is now with the E.W. Laboratory of John Brown and Co. (Clydebank) Ltd.

Mr. **J. Russell Taylor,** T.D., at present Director of Development and Research of the Igranic Electrical Co., Ltd., is relinquishing his departmental responsibilities on 1st January to become consultant to the Managing Director on all matters connected with development, research, and production.

Mr. **T. R. Twigger** has been appointed to the Board of The British Piston Ring Co., Ltd., and those of its subsidiaries Bricmovo Ltd. and British Aero Components Ltd.

Dr. **T. Vér** is now with the Research Laboratory of the Cincinnatti Milling Machine Company.

Mr. **A. J. Williamson** has resigned his position as President and Director of the Tube Reducing Corporation of Wallington, New Jersey, U.S.A.

Mr. **A. B. Woodhead** has joined Personnel Administration Ltd.

Mr. **C. B. Wright** has been appointed Technical Sales Manager of Davy and United Roll Foundry Ltd.

Obituary

Dr. **Richard William Bailey,** D.SC., F.R.S. (*elected* 1937) of Manchester, on 4th September, 1957.

Dr. **Paul Dyer Merica** (*elected* 1921), of New York, on 20th October, 1957.

Mr. **Edgar Alleyne Reynolds** (*elected* 1919) of Scunthorpe.

Mr. **Frederick John Skinner** (*elected* 1924), of East Molesey, on 9th December, 1956.

Com. **Amador Gonzalez Soto** (*elected* 1928), of Madrid, in March, 1957.

Dr. **Vladimir Zednik** (*elected* 1947), of Moravská Ostrava, Czechoslovakia, in 1953.

IRON AND STEEL ENGINEERS GROUP

Meeting on Water Pollution

The Iron and Steel Engineers Group of the Institute is holding a two-day meeting on "Water Pollution in the Iron and Steel Industry" on Wednesday and Thursday, **11th and 12th December**. The meeting will take place at Church House, Great Smith Street, London, S.W.1.

The proceeding will be opened by Mr. W. M. Larke, Chairman of the Group, at 9.45 a.m. on 11th December. The technical sessions, details of which were given in the November *Journal* (p. 242), are from 10.0 a.m. to 4.30 p.m. on both days. Buffet luncheons can be obtained at Church House.

POWDER METALLURGY JOINT GROUP

Inaugural Meeting

The recently formed Powder Metallurgy Joint Group of The Iron and Steel Institute and the Institute of Metals is holding its inaugural meeting at Church House, Great Smith Street, London, S.W.1., on Wednesday, **4th December**, at 10.30 a.m.

Tickets are not required for the meeting, and visitors will be welcome. Details of membership of the Group, which is restricted to members of the two Institutes, may be obtained from the Secretary of the Group, Lt.-Col. S. C. Guillan, 17 Belgrave Square, London, S.W.1.

CONTRIBUTOR TO THE JOURNAL

D. A. Leak, B.SC.—Metallurgy (General) Division, British Iron and Steel Research Association, Sheffield.

Mr. Leak was educated at Thames Valley Grammar School, Twickenham. After service with the Royal Electrical and Mechanical Engineers, he entered Queen Mary College, London University, and obtained a B.Sc. General degree in physics and mathematics in 1952. He worked for a short time with E.M.I. Ltd. on a Ministry of Supply project after graduating, and joined B.I.S.R.A. in 1953. Since then he has been working with his brother, Dr. G. M. Leak, on an investigation into the influence of impurities on the magnetic properties of silicon–iron.

(*Mr. Leak was joint author of a paper published in the November issue of the Journal*).

D. A. Leak

G. Lonie

G. Lonie, A.R.C.S.T.—Technical and Research Department, Consett Iron Company, Ltd.

Mr. Lonie was educated at Tonbridge School and the Royal College of Science and Technology, Glasgow, where in 1954 he was awarded an Associateship in Metallurgy, with Honours. He had spent a year in the Research Department of Colvilles Ltd., before entering the College.

In 1954 Mr. Lonie joined the British Coke Research Association, and was attached to the Metallurgy Department of his old college. He joined his present company in March, 1957.

AFFILIATED LOCAL SOCIETIES

Newport and District Metallurgical Society

The following Committee and officers have been elected for the 1957–1958 session:

President

G. H. Latham, LL.D., J.P.

Hon. Secretary	*Hon. Treasurer*
T. G. Grey-Davies	C. J. Morgan, A.I.M.

Committee

C. I. F. Mackay (*Chairman*)	H. T. Roberts, A.I.M.
H. R. Davies, A.I.M.	H. H. Stanley, A.MET., F.I.M.
B. Jones, D.SC., F.R.I.C.	G. L. Jones, A.I.M.
R. A. Owen-Barnett, M.SC.	C. B. Pennington
J. E. Wells, B.SC., F.R.I.C., A.MET.	

History of the Iron and Steel Industry

The "History of the Iron and Steel Industry" by Dr. H. R. Schubert (see November *Journal*, p. 239) is being published early in December. Members are reminded that copies at the special rate of 50*s*. can be obtained on application to the Secretary, using the form circulated with the November issue.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

B.I.S.R.A. Survey for 1957

The 1957 edition of "B.I.S.R.A. Survey" contains, as in previous years, details of research projects which have reached an interesting stage of development.

A substantial proportion of the contents deals with developments in various aspects of process control, reflecting current interest in the application of automation to raise productivity. The projects described include the mechanization of forging methods, the automatic control of heavy plant, such as rolling mills, using a translator system, optical callipers for measuring dimensions of hot materials, and micro-analysis using electron probes (the Castaing technique).

Progress in iron and steelmaking techniques benefits from researches such as an investigation into the characteristics of the ideal stack refractory material, the self-lining blast-furnace (see *J. Iron Steel Inst.*, 1957, vol. 187, pp. 32–45), the production of low-sulphur iron, and the effect of the increased availability of tonnage oxygen. Another 'live topic' under investigation is the emission of fume in steelmaking.

This handsome 16-page survey can be obtained from B.I.S.R.A., 11 Park Lane, London W.1.

Load and Torque in Flat Rolling

A book entitled "Calculation of Load and Torque in Hot Flat Rolling" is shortly to be published by B.I.S.R.A.; it should be of considerable interest to all engaged in hot rolling. The results of five years' work with the Cam Plastometer at B.I.S.R.A.'s Sheffield laboratories are reported; it was used for determining the stress/strain curves for steel in compression at temperatures up to 1200° C and over a range of deformation speeds usually met in working practice.

The book will describe the method and contain the data for calculating hot load and torque for twelve steels. Ninety-six large graphs will give the necessary 'mean yield strength' data for these steels at 900°, 1000°, 1100°, and 1200° C, and over a range of strain rates and reductions covering all industrial requirements. Other graphs will be included from which the strain-rate and geometrical functions associated with any particular rolling operation can rapidly be calculated.

This is the first time that such a detailed and informative body of exact information on the subject has been published, and it should prove a reference work of great value to the rolling industry. The book, which is being produced on robust lines with facilities for inserting further data sheets as these become available, will probably be priced at about £3 0*s*. 0*d*. It is hoped it will be published in December; further details will be issued in due course.

INSTITUTION OF METALLURGISTS

New Registrar-Secretary

The Council of the Institution of Metallurgists announces that Dr. A. D. Merriman, G.C., M.A., M.ED., C.I.MECH.E., F.R.S.E., who has been Registrar-Secretary for the past ten years, will retire on 31st December, 1957. During Dr. Merriman's tenure of office the Institution has grown from a membership of less than 1600 to more than 3500, and in a recent appreciation of his services by Professor F. C. Thompson (Manchester University), a Past-President, tribute was paid to his part in firmly establishing the Institution as the professional organization of metallurgists.

Dr. Merriman will be succeeded on 1st January by Mr. R. G. S. Ludlam, B.A., who has been Assistant Registrar in the University of Leeds since 1951. Mr. Ludlam, who is 36, was educated at schools in South Africa and Southern Rhodesia, followed by King Edward VII School at Sheffield. He graduated in the University of Sheffield with First Class Honours in English Language and Literature in 1948 and was awarded the University English Prize and the Gibbons Prize in that year. During the war he served in the Royal Air Force and before joining the staff of the University of Leeds he spent two years in the Department of the Clerk to the London County Council.

INSTITUTE OF METALS

Symposium on Point Defects

The Metal Physics Committee of The Institute of Metals has organized a Symposium on "Vacancies and Other Point Defects in Metals and Alloys", to be held at the Cockcroft Hall, Harwell, near Didcot, Berks., on Monday and Tuesday, **9th and 10th December.**

On the first day there will be a visit to the Atomic Energy Research Establishment, and also to the reactor DIDO, in the afternoon. In the evening Professor R. W. K. Honeycombe, of the University of Sheffield, will deliver an introductory lecture for the benefit of non-specialists. The symposium itself will be held on the second day; the discussion will be based on a series of six papers, preprints of which will be available in advance to those registering for the meeting.

The symposium is open to non-members of The Institute of Metals as well as members, but attendance must be by registration only. A basic registration fee of £1 5*s*. 0*d*. covers the cost of advance copies of the papers; meals and transport to Oxford and Didcot will be provided at additional cost for those who desire them. Details of the meeting and registration forms may be obtained from the Secretary, The Institute of Metals, 17 Belgrave Square, London, S.W.1.

Meeting on Fuel Efficiency

The Metallurgical Engineering Committee is arranging an all-day formal discussion on "Fuel Efficiency in the Melting and Thermal Treatment of Metals", to be held at the College of Technology, Gosta Green, Birmingham, on Wednesday, **26th February, 1958**. Visitors will be welcome and tickets are not required.

BRITISH NUCLEAR ENERGY CONFERENCE

Symposium on Nuclear Energy

The Institution of Chemical Engineers is holding an all-day Symposium on Nuclear Energy at the Hoare Memorial Hall, Church House, Great Smith Street, London, S.W.1, on 21st January, 1958. Papers to be presented include "The Commissioning and Operation of Calder Hall," by H. G. Davey, "Fission Product Disposal," by K. Saddington, "The Marcoule Plutonium Plant," by R. Galley, etc.

Further details and registration forms may be obtained from the Secretary, The Institution of Chemical Engineers, 16 Belgrave Square, London, S.W.1.

EDUCATION

Sandwich Course in Metallurgical Engineering

The redevelopment of the British iron and steel industry has created a need for engineers of a special kind; they must be able to cope with a stream of problems, most of which have no text-book solution. Thus works experience is of very great importance, and so a new approach to the training of engineers to work in steelworks became imperative.

The Training Committee of the British Iron and Steel Federation decided that the best answer would be to provide a sandwich course leading to the Diploma in Technology; the course would combine the essential works experience with a high standard of fundamental knowledge. After consultations with companies, research associations, and colleges, Loughborough College of Technology was invited to make arrangements for the course on the lines laid down after careful discussion. Recognition for the course is now being sought and the first set of students will enrol in January, 1958.

NEWS OF SCIENCE AND INDUSTRY

125 Years of Baileys

The 125th anniversary of the foundation of Sir W. H. Bailey and Co., Ltd., Patricroft, Manchester, was celebrated recently.

The company was founded at Salford in 1832 by John Bailey. Production was at first restricted to turret clocks, but gradually a whole range of engineering products was added, including gauges, testers, and boiler fittings. On the retirement of the founder in 1865 he was succeeded by his son (Sir) William Henry Bailey, who carried out a policy of energetic expansion. Sir William Bailey's interests were many and varied; his enthusiasm and energy in carrying out the Manchester Ship Canal project led to his being knighted by Queen Victoria on board the first ship to pass through the Canal. He was a Member of The Iron and Steel Institute for many years.

The wide range of Bailey products was amply illustrated by their famous catalogues, "Bailey's Useful Inventions for Engineers," the last of which appeared before the First World War; a few of the items described, taken at random, are a hydraulic organ blower, a water-motor pumping engine, a patent beer raiser and aerator, lightning conductors, hand lathes, gyrometers, and a great variety of valves. The original Salford works was destroyed by enemy action in 1940, but work went on at the surviving foundry at Patricroft. Since the war the works has been considerably expanded, and now over 300 people are employed. Production is now centred chiefly on the well-known Bailey turnstiles and an extensive series of valves.

Some details of the history of the company preface the 125th anniversary catalogue, entitled "A Guide to Valve Practice," the first to be issued since 1908.

New Electrode for Welding Mild Steel

A new type of manual welding electrode which combines the good welding properties of the B.S.1719 Class 2 electrodes with the good mechanical properties and X-ray sound weld metal of the low-hydrogen Class 6 electrodes has been introduced by Quasi-Arc Ltd. This electrode, "Ferromax," has a rutile covering containing iron powder, which deposits a weld metal which is very low in hydrogen content. As a result, a better shape of fillet weld is obtained compared with the standard low-hydrogen electrodes, there is considerable reduction in fumes, and welding speeds are greater.

Royal Visit to Hadfields

H.R.H. Prince Philip, Duke of Edinburgh, visited the East Hecla Works, Sheffield, of Hadfields Ltd., on 24th October. The Royal visitor saw during his tour No. 1 machine shop, the pattern shop, the radiographic laboratory, the foundry, the precision foundry, the light forge, the heavy forge and heat-treatment shop, the welding laboratory, and the shell moulding division.

Teaching of Corrosion Science

A discussion on the teaching of the science of metallic corrosion has been arranged by the Corrosion Group of the Society of Chemical Indsutry and will be held at 14 Belgrave Square, London, S.W.1, on Wednesday, 18th December, from 2.30 to 4.00 p.m. and from 4.45 to 6.30 p.m.

The first part of the discussion, opened by Mr. L. W. Stubbs, will relate to the facilities at present available in universities and technical colleges; the second part, opened by Dr. T. P. Hoar, will be concerned with future developments.

The meeting is open to all and no tickets are required.

Industrial Publications

Two new furnaces for handling aluminium are described in technical information circulars on "The Birlec–Morgan Electric Die-casting Furnace Type EDF Mk. I" and "The Morgan Basin Tilting Crucible Furnace Type BT Mk. II," available from Morgan Crucible Co., Ltd., Battersea Church Road, London, S.W.11.

"*Automation and Electroplating*" is the title of a brochure published by Electro-Chemical Engineering Co., Ltd., Sheerwater, Woking, Surrey.

A technical information booklet on Brimor high-temperature cements is obtainable from the Morgan Crucible Co., Ltd., Battersea Church Road, London, S.W.11.

John Williams and Sons (Cardiff) Ltd., East Moors Road, Cardiff, have recently published a brochure entitled "*Sound and Solid Castings*," which describes the range of castings in grey, alloy, and s.g. irons which they manufacture.

An employee publication of the United Steel Companies Ltd. called "*Savings and Investment*" describes in simple terms the principles of shareholding and some methods of investment. It emphasizes that industry needs the savings of its workpeople to finance its future.

Brochure no. 574, "*Modern Dust Collection*," is a new publication obtainable from the Visco Engineering Co., Ltd., Stafford Road, Croydon, Surrey.

A laboratory resistance-heated vacuum furnace (model

2904) is described in a recent brochure from Wild-Barfield Electric Furnaces Ltd., Elecfurn Works, Otterspool Way, Watford.

Carter Gears Ltd., Thornbury Road, Bradford 3, have published a booklet describing their " A " type hydraulic infinitely variable speed gears and auxiliary equipment.

Publication No. 38 of Keith Blackman Ltd., Mill Mead Road, London, N.17, deals with the " Tornado " four-way welding fume exhaust unit.

The Holmes–Catco Process, a low-temperature oxidation process for converting undesirable fumes, vapours, and gases to odourfree gases, is described in Publication No. 70 of W.C. Holmes and Co., Ltd., Turnbridge, Huddersfield.

" Gas Cleaning Plant " is the title of the latest brochure issued by Head Wrightson Iron and Steel Works Engineering Ltd., Teesdale Iron Works, Thornaby-on-Tees.

DIARY

2nd Dec.—CLEVELAND INSTITUTION OF ENGINEERS—" The Use of Sinter on the Blast-furnaces at Dorman Long (Steel) Ltd.," by A. T. Ledgard and Dr. P. K. Gledhill—Cleveland Scientific and Technical Institution, 6.30 P.M.

2nd Dec.—EAST MIDLANDS METALLURGICAL SOCIETY—" Recent Developments in Plain Bearings," by M. J. Neale—College of Art, Green Lane, Derby, 7.30 P.M.

3rd Dec.—NEWPORT AND DISTRICT METALLURGICAL SOCIETY—" Metallurgical Problems of Guided Missiles and High Speed Aircraft," by H. Brooks—Whitehead Institute, Cardiff Road, Newport, 7 P.M.

3rd Dec.—SHEFFIELD METALLURGICAL ASSOCIATION—" The Physical Chemistry of Chemical Analysis," by C. E. A. Shanahan—B.I.S.R.A., Hoyle Street, Sheffield, 7 P.M.

4th Dec.—MANCHESTER METALLURGICAL SOCIETY—" The Metallurgical Applications of High-resolution Electron Microscopy," by Dr. J. Nutting—Central Library, Manchester, 6.30 P.M.

4th Dec.—POWDER METALLURGY JOINT GROUP (The Iron and Steel Institute and The Institute of Metals)—Inaugural Meeting—Church House, Westminster, London, S.W.1, 10.30 A.M.

5th Dec.—BRITISH NUCLEAR ENERGY CONFERENCE—" Advanced Types of Power Reactor," by Dr. J. V. Dunworth, C.B.E.—Institution of Electrical Engineers, 5.30 P.M.

5th Dec.—INSTITUTE OF METALS AND SOCIETY OF CHEMICAL INDUSTRY—" Design and Operation of Waste Heat Boilers in the Chemical Industry," by Capt. W. Gregson—The University, Woodland Road, Bristol, 6.30 P.M.

5th Dec.—LEEDS METALLURGICAL SOCIETY—" Some Uncommon Failures of Engineering Plant in Service," by G. A. Cottell—Chemistry Wing, The University, Leeds, 7.15 P.M.

6th Dec.—SHEFFIELD METALLURGICAL ASSOCIATION—Dinner Dance—Royal Victoria Hotel, Sheffield.

9th–10th Dec.—INSTITUTE OF METALS—Symposium on " Vacancies and Other Point Defects in Metals and Alloys "—Atomic Energy Research Establishment, Harwell.

10th Dec.—INSTITUTE OF METALS (South Wales Local Section)—" Hardness of Metals," by Dr. D. Tabor—Department of Metallurgy, University College, Swansea, 6.30 P.M.

10th Dec.—SHEFFIELD METALLURGICAL ASSOCIATION—" Some Everyday Refractory Problems," by Dr. P. W. Clark—B.I.S.R.A., Hoyle Street, Sheffield, 7 P.M.

11th Dec.—INSTITUTE OF METALS (London Local Section) and INSTITUTE OF BRITISH FOUNDRYMEN (London Branch)—" Properties of Castings," by F. Hudson—Constitutional Club, Northumberland Avenue, London, W.C.2, 7.30 P.M.

11th Dec.—NEWCOMEN SOCIETY—" Abraham Darby (I and II) and the Coal–Iron Industry," by Dr. R. A. Mott.

11th Dec.—NORTH EAST METALLURGICAL SOCIETY—" Stress-corrosion Cracking," by Dr. T. P. Hoar—Cleveland Scientific and Technical Institution, Middlesborough, 7.15 P.M.

11th Dec.—NORTH WALES METALLURGICAL SOCIETY—" Statistical Method in the Iron and Steel Industry," by G. Millington—Flintshire Technical College, Connah's Quay, near Chester, 7.15 P.M.

11th and 12th Dec.—THE IRON AND STEEL INSTITUTE (Engineers Group)—Meeting on Water Pollution in the Iron and Stee Industry—Church House, Westminster, London, W.1.

12th Dec.—SOCIETY OF INSTRUMENT TECHNOLOGY—" Calibration of Thermocouples," by W. I. Morrow—Cleveland Scientific and Technical Institution, 7.30 P.M.

12th Dec.—STAFFORDSHIRE IRON AND STEEL INSTITUTE—Film Evening—Station Hotel, Dudley, 7.30 P.M.

14th Dec.—SWANSEA AND DISTRICT METALLURGICAL SOCIETY—" Metallurgical Aspects in the Manufacture of Cold Reduced Tinplate," by I. Williams—Central Library, Swansea, 6.30 P.M.

17th Dec.—SHEFFIELD METALLURGICAL ASSOCIATION—Film Night—B.I.S.R.A., Hoyle Street, Sheffield, 7 P.M.

18th Dec.—SOCIETY OF CHEMICAL INDUSTRY (Corrosion Group)—" Teaching Corrosion Science," by L. W. Stubbs and Dr. T. P. Hoar—14 Belgrave Square, London, S.W.1, 2.30 P.M.

20th Dec.—WEST OF SCOTLAND IRON AND STEEL INSTITUTE—" Electric Melting Furnaces," by F. S. Leigh—39 Elmbank Crescent, Glasgow, 6.45 P.M.

26th–31st Dec.—AMERICAN ASSOCIATION FOR THE ADVANCEMENT OF SCIENCE—Annual Meeting—Indianopolis, Ind., U.S.A.

CO-OPERATIVE TRANSLATION SERVICE

The following translations are now available, in addition to those given on pp. 243 and 280 of the November, 1957, issue of the *Journal*:

571. MAROT: " Determination of Nitrogen in Cast Iron." *Rev. Mét.*, 1956, vol. 53, No. 9, pp. 689–700. (£3 10*s*. 0*d*.)

577. CHATEL: " Study of Ridge Formation in Hot Dip Tinning." *Métaux–Corrosion–Industries*, 1955, vol. 30, Feb., pp. 78–87. (£3 5*s*. 0*d*.)

582. AMMANN and HINNUBER: " Development of the Hard-Metal Alloys in Germany." *Stahl u. Eisen*, 1951, vol. 71, Oct. 11th, pp. 1081–1090. (£5 0*s*. 0*d*.)

600. MOREAU: " Effect of Furnace Atmosphere on Scaling of Billets." *Centre Doc. Sidér., Circ. Inform. Tech.*, 1956, No. 10, pp. 1973–1982. (£1 10*s*. 0*d*.).

613. BJERKERUD: " Apparatus for Determination of Hydrogen in Steel " and " Sampling when Determining Hydrogen in Steel." *Jernkontorets Ann.*, 1957, vol. 141, pp. 90–99. (£2 10*s*. 0*d*.)

614. ABE and MIWA: " Extensometer for Skin-pass Mills." *Tetsu to Hagane*, 1956, vol. 42, No. 3, pp. 342–343. (£1 15*s*. 0*d*.).

615. LEITNER: " Experiences with the Application of Basic Brick-work in O.H. Furnaces." *Neue Hütte*, 1957, Feb./Mar., No. 2/3, pp. 142–154. (£5 5*s*. 0*d*.)

619. SEPTIER *et al.*: " Improvement of the Output of a Basic Converter Steel Plant by using Large Bottom Converters." *Centre Doc. Sidér., Circ. Inform. Tech.*, 1956, vol. 13, No. 7, pp. 1415–1444. (£4 0*s*. 0*d*.)

632. TSELIKOV: " Influence of Outer Zones on Widening and Distribution of Speeds and Stresses across the Width of a Strip during Rolling." From " Problems of Metallurgy," pp. 392–406: 1953, Moscow, U.S.S.R. Acad. Sci. (£3 10*s*. 0*d*.)

633. LEFRANC: " The Maprodent Process of Precision Casting and Application in Aeronautical Industries." *La Technique Moderne*, 1955, vol. 47, Sept., pp. 390–392. (£1 10*s*. 0*d*.)

638. FLAMENT: " Influence of Small Quantities of Aluminium on Certain Mechanical Properties of Nitrogen-bearing Ferrite." *Rev. Mét.*, 1957, vol. 54, July, pp. 537–554. (£6 15*s*. 0*d*.)

646. KOCHENDORFER and SCHOLL: " The Tendency of Steels to Brittle Fracture." *Stahl u. Eisen*, 1957, vol. 77, July 25th, pp. 1006–1018. (£6 15*s*. 0*d*.)

648. BOGDANDY and SCHMOLKE: " Production of Finest Ore Agglomerates and their Behaviour in the Blast Furnace." *Stahl u. Eisen*, 1957, vol. 77, May 30th, pp. 685–692. (£4 15*s*. 0*d*.)

649. BRANDES and WENDEBORN: " Sinter Cooling: an Essential Element of Modern Plant." *Stahl u. Eisen*, 1957, vol. 77, May 30th, pp. 693–700. (£5 5*s*. 0*d*.)

651. SCHULL and ROCHROHR: " The Basic Hot-blast Cupola as a Pre-melting Unit in the Open-hearth Steel Works." *Stahl u. Eisen*, 1956, vol. 76, Jan. 12th, pp. 1–13. (£6 0*s*. 0*d*.)

654. DECKER: " Chemical Equilibria in the Basic Bessemer Converter." *Arch. Eisenhüttenwesen*, 1957, vol. 28, Feb., pp. 57–64. (£3 15*s*. 0*d*.)

658. MAKAEV *et al.*: " The New Wheel Rolling Plant in Use at the Nizhne-Tagilskoi Steel Combine." *Stal*, 1957, July, pp. 616–621. (£3 10*s*. 0*d*.)

671. KÖRÖS: " Production of Rolling Mill Rolls from Magnesium-treated Spheroidal Graphite Cast Iron." *Freibergische Forschung*, 1956, Section *B*, No. 7, May, pp. 80–108. (£4 0*s*. 0*d*.)

678. SCHULZ: " The Coating of Metals with Synthetic Materials." *Schweizer Archiv*, 1956, June, pp. 178–182. (£2 10*s*. 0*d*.)

681. RASPOPOV *et al.*: " Smelting of Fluxed Sinter made from Krivoi Rog Iron Ores." *Stal*. 1957, Feb., p. 99. (£2 0*s*. 0*d*.)

686. VON DER DUNK: " Corrosion Resistance of Aluminized Soft Steel." *Stahl u. Eisen*, 1957, vol. 77, Mar. 7th, pp. 294–296. (£1 10*s*. 0*d*.)

690. CHERNOV *et al.*: " Location of Blast-furnace Mixing Valve." *Stal*, 1957, May, pp. 389–391. (£1 15*s*. 0*d*.)

691 MOISEYEVICH: " Hot-cooling of Open-hearth Furnaces." *Stal*, 1957, July, p. 658. (£2. 0*s*. 0*d*.)

717. PROBEDIN and PRICHODKO: " Machines for the Marking of Blooms and Slabs." *Stal*, 1957, Apr., pp. 333–340. (£3 10*s*. 0*d*.)

731. BRON and LANDE: " Alumina–Chrome Refractories for Lining Steelworks Ladles." *Metallurg*, 1957, Sept., pp. 17–18. (£1 0*s*. 0*d*.)

Note.—When two copies are required, there will be 50% reduction for the second copy. Six copies can be bought for the price of two, and twelve copies for the price of three.

ABSTRACTS OF CURRENT LITERATURE AND BOOK NOTICES

Iron and Steel Manufacture and Related Subjects

CONTENTS

	PAGE
Mineral Resources	349
Ores—Mining and Treatment	349
Fuel—Preparation, Properties, and Uses	350
Refractory Materials	350
Iron and Steel, General	351
Blast-Furnace Practice and Production of Pig Iron	352
Properties, Treatment and Use of Slag	353
Direct Processes	353
Production of Steel	353
Production of Ferro-Alloys	355
Foundry Practice	356
Vacuum Metallurgy	358
Reheating Furnaces and Soaking Pits	359
Heat-Treatment and Heat-Treatment Furnaces	359
Forging, Stamping, Drawing, and Pressing	361
Rolling Mill Practice	361
Machinery for Iron and Steel Plant	363
Welding and Flame-Cutting	363
Machining and Machinability	365
Cleaning and Pickling	365
Protective Coatings	365
Powder Metallurgy	367
Ferrites, Cermets, and Carbides	367
Properties and Tests	367
Metallography	374
Corrosion	378
Analysis	380
Industrial Uses and Applications	381
Historical	381
Economics and Statistics	381
Miscellaneous	382
Book Notices	382
New Publications	383

MINERAL RESOURCES

Bibliography of North American Geology, 1954. R. R. King. (*U.S. Geol. Survey Bull.*, 1054, 1957, pp. 484). Economic minerals are included.

Tungsten Deposits in the Fairbanks District, Alaska. F. M. Byers, jun. (*U.S. Geol. Survey Bull.*, 1024–I, 1957, 179–216).

A Reconnaissance Study of the Beach Sands of Puerto Rico. R. B. Guillou and J. J. Glass. (*U.S. Geol. Survey Bull.*, 1042–I,, 1957, 273–303). Magnetite is abundant in some places, its distribution is tabulated.

ORES—MINING AND TREATMENT

Deformation Measurements in the Lorraine Iron Mines. E. Tincelin and P. Sinou. (*Mine Quarry Eng.*, 1957, **23,** July, 299–305; Aug., 350–354). Effects of widening single drifts and driving parallel entries are described, and the correlation between surface subsidence and underground deformation and the measurement of expansion towards mining cavities and the use of roof bolts are discussed.

Mining East Texas Iron Ore. V. F. Malone. (*Min. Eng.*, 1957, **9,** May, 524–527). Details are given of the ore mining operations of Lone Star Steel Co. near Daingerfield. Exploration is not complete, but 75 years' supplies have been proved, 850,000 tons of concentrates are produced annually from 4 million tons of crude ore; the blast furnace produces 1200 tons of iron per day. Limonite and siderite ores, in seams 1–3 ft thick, are extracted by open-cast methods (drag-line mainly, but drilling and blasting for hard ore) from an overburden, the economic limit of which is 1 cu. yd. per ton of ore. The ores are blended in their type, crushed, washed and screened. The coarse limonite is roasted to eliminate water; coarse siderite is calcined to eliminate CO_2 and reduce sulphur; and fines are sintered. Some sulphur is eliminated from the siderite ores by weathering, when sulphate is produced.

Narvik Conveyors Sort, Mix and Load Sweden's Ores. (*Eng. Min. J.*, 1957, **158,** June, 92–94). An illustrated account is given.

The Study of Titanium–Iron Spinel from a Titaniferous Iron Deposit in the Province of Quebec. (*Canad. Mines Tech. Surveys Research and Special Projects*, (*July–Dec.*, 1956), 1957, Feb. 8, 15–16).

Indian Manganese Ores and their Thermal Beneficiation for Ferromanganese Manufacture. B. D. Nijhawan. (*Symposium on Mineral Beneficiation and Extractive Metallurgical Techniques, Council Sci. Ind. Res. Jemshedpur, India*, 1957, Feb. 8–11, pp. 20). Compositions and locations are given and beneficiation processes are reviewed. A pilot-plant scheme using a rotary furnace is outlined and a low-shaft furnace project to use non-coking coals is also referred to.

Experiments on the Calcination of Limestone on a Sintering Machine. T. A. Kramnik and A. A. Sorokin. (*Stal'*, 1956, (9), 774–776). [In Russian]. It is claimed that laboratory and full-scale experiments have confirmed that even if the calcination of limestone is carried out on the strand the loss of machine time is more than compensated for by the greater sintering speed obtained with lime additions.—S. K.

Quality Control in Selective Mining of Magnesite. C. Martin and H. P. Willard. (*Mining Eng.*, 1957, **9,** Apr., 425–427).

The Importance of Preparation of Home Ore to the West-German Iron and Steel Industry. E. Böhne. (*Stahl u. Eisen*, 1957, **77,** May 2, 549–552). The author lists the various sources of home ore in Germany. The iron content ranges from 21 to 34%. In 1955, 15·7 million tons of home ore having an average iron content of 26·9% was used. Half of it was subjected to ore preparation attaining an iron content of 38% and a yield of 81%. The various methods of preparation are outlined.—T. G.

Preconcentration of Carbonate-type Niobium Ores by Calcination and Washing. (*Canad. Mines Tech. Surveys, Research and Special Projects Report* (*July-Dec.* 1956), 1957, Feb. 8, 1–3). A laboratory study.

The Role of National Metallurgical Laboratory in the Development of Mineral Beneficiation and Extractive Metallurgical Techniques in the Second Five Year Plan. B. R. Nijhawan. (*Symposium on Mineral Beneficiation and Extractive Metallurgical Techniques, Council Sci. Ind. Res. Jemshedpur, India*, 1957, Feb. 8–11, pp. 49). Indian mineral resources are reviewed in the light of the 5-year plan and work in the laboratory is outlined on Mn, chromite, graphite, pyrites, etc. and on pilot-plant work on electrolytic Mn and MnO_2. A low carbon ferrochrome process in which chromite is reduced by Fe–Si and Al is also described and the production of Al–Si by aluminothermic reduction of quartz. Other non-

ferrous topics are mentioned. Work is planned on beneficiation of Mn and Cr ores and pilot-plant studies on refractories.

Electrical Equipment for [Ore] Processing Plants. C. B. Risler and W. E. Thomas. (*Min. Eng.*, 1957, **9**, May, 529–537). A general review is made of equipment for ore-treating plants; since most of the duties are continuous, squirrel-cage, wound rotor or synchronous a.c. motors are generally used. General drive and control characteristics are stated, and detailed discussion is given for gyratory and cone crushers, roll crushers, jaw crushers, hammer mills, rod and ball mills, flotation cells, filters, screens, conveyors, cranes, pumps, compressors and blowers, and distribution systems.—K. E. J.

Sinter Bed Ignition. H. Bates. (*J. Iron Steel Inst.*, 1957, **187**, Dec., 310–314). [This issue.]

Sintering of Iron Ore Fines with Particular Reference to the Raw Materials for the Bhilai Steel Plant. G. P. Mathur, G. V. Subramanya and P. I. A. Narayanan. (*Symposium on Mineral Beneficiation and Extractive Metallurgical Techniques, Council Sci. Ind. Res. Jemshedpur, India*, 1957, Feb. 8–11, pp. 17). Coke, 6·1%, was optimal for good sinter production in a 3-layer bed, a simplified method with 5·3% coke was equally satisfactory. Moisture affects sintering rate but not sinter properties. Basicity for shortest time of sintering was 0·8, time and sinter strength increase up to 0·92. Laterite increased sintering time but was without effect on strength. The sinters produced were strongly magnetic. Various tests for sinter quality are reported.

An Improved Method of Sintering High-Volatile Iron Ores. R. F. Jennings. (*Iron Coal Trades Rev.*, 1957, **175**, July 5, 15–19). A two-layer process is suggested and the thermal balance compared with that of the conventional process. Materials balances are given diagrammatically and advantages and limitations are set out.

Fluxed Sinter with an Elevated Magnesia Content. A. I. Gamayurov and A. G. Neyasov. (*Stal'*, 1957, (1), 20–24). [In Russian]. On the basis of previous work the suggestion had been made that the magnesia content in sinter should be increased to improve its strength, and reducibility. This was confirmed in the work described, in which sinters were made in a 400-mm box and their properties studied. Magnesia was found to have a big effect on the properties: for uniform properties the ratio (CaO + MgO) : (SiO_2 + Al_2O_3) and MgO : (CaO + MgO) or the ratio CaO : SiO_2 and MgO : (CaO + MgO) must be kept constant. To find the optimal MgO-content, blast-furnace tests with sinters containing over 3% MgO are urged.—S. K.

Work on the Automatic Speed Control of a Sintering Machine in Relation to the Completion of the Process. D. G. Khokhlov, Yu. A. Gyrdymov and M. M. Gordon. (*Stal'*, 1957, (6), 481–488). The waste gas temperature difference in the last and last-but-one wind boxes was used to control the rate of the machines. Measurements were made by two differential thermocouples. Improved quality, increased productivity and better working conditions and improved sampling are claimed. To complete automatic control, a sinter-mix maker and control of humidification are needed.

Dust Removal From Gases Evolved on Sinter Belts. H. Krumnikl. (*Hutník*, 1957, **7**, (3), 74–81). [In Czech]. The mode of operation and effectiveness of venturi-type, cyclone-type and other dust-separators are discussed. Multi-cyclone types are considered most appropriate under sintering conditions prevailing in Czechoslovak plants making blast-furnace agglomerates.—P. F.

FUEL—PREPARATION, PROPERTIES, AND USES

Coal Geology of the White Oak Quadrangle, Magoffin and Morgan Counties, Kentucky. W. L. Adkinson. (*U.S. Geol. Survey Bull.*, 1047–A, 1957, 1–21).

Analysis of the Inorganic Constituents of Coal by Photometric Methods. W. J. S. Pringle. (*Fuel*, 1957, **36**, July, 257–276). After wet oxidation and determination of SiO_2, Al, Fe, Ti, P and Mn are determined directly and Ca, Mg, Na and K are isolated and the precipitates examined photometrically.

Results of Research on the Improvement of Flotation, and their Applicability in Practice. P. Belugou. (*Glückauf*, 1957, **93**, 16 Mar., 313–321). This is a German translation of an abridged French paper. Factors influencing coal flotation results were studied under laboratory conditions; and the results obtained were tested in several French flotation plants. The trials showed indicated that improvements were possible in many cases.—L. D. H.

The Effectiveness of Modern Coal Quality Assessments. A. Kozina. (*Sborník Vědeckých Prací Vysoké Školy Báňské v Ostravé*, 1956, 2, (2), 3–18. [In Czech]. Modern methods, particularly the author's development of the use of the distribution curve of flotation fractions, i.e. specific weight distributions are discussed, and practical applications are considered.—P. F.

Assessment of Coals for Industrial Use: Some Current Investigations into the Properties and Characterization of Australian Coals. H. R. Brown. (*J. Inst. Fuel*, 1957, **30**, Mar., 137–155). The work presented includes carbonization studies on the following subjects:—differential thermogravimetric analysis: effect of heating rates on decomposition: thickness of the plastic layer.—D. L. C. P.

Coke-Oven Technology in the U.S.S.R. P. Bezouhoff. (*Coke Gas*, 1957, **19**, July, 288–291, 295). A general review of production, coal blending ovens and their performance, and of coke quality is given.

Experiments Concerning the Effect of Coking Pressure in Full Scale Operation of Coke Oven Plant. J. L. Lambert, W. J. Lancucki-Pater and G. W. Lee. (*J. Inst. Fuel*, 1957, **30**, July 362–386; *Gas World*, 1957, **146**, July 6, *Coking Section*, 5–20. Movable wall test oven tests suggest that coal crushed to 80% passing $\frac{1}{8}$ in. mesh with over 7% moisture should show a max. pressure below 2 lb/in² to avoid damage and this figure has been confirmed by tests on a full scale plant.

Making Coke Without Coke Ovens. G. V. Speranskaya and Yu. M. Skvortsov. (*Iron Coal Trades Rev.*, 1957, **174**, June 14, 1383–1384; from *Koks i Khimiya*, 1956, (7)). Feebly-caking Russian coals were concentrated (for ash removal) and a process used in which the coal was subjected to a heating process, compressed, and the product sintered in a continuous plant. Properties of the coals and metallurgical coke briquettes obtained are outlined.

Automatic Regulation of the Operation of Gas-Mixing Installations. A. P. Kopelovich and S. Ya. Peliks (*Stal'*, 1956, (11), 1036–1042). [In Russian]. A critical survey is made of various gas-mixing systems, with special reference to the automatic regulation of coke-oven/blast-furnace gas-mixtures to supply a product of constant calorific value. Best results are attributed to systems based on paired valves in the respective mains.—S. K.

The Radiation from Flames in a Small-scale Oil-fired Furnace. D. K. Holliday, M. W. Thring. (*J. Inst. Fuel*, 1957, **30**, Mar., 127–136). Experiments were made on a small experimental furnace to investigate the effect of liquid fuel composition on flame radiation, and to provide information on the relation between large and small scale furnace investigations. It is concluded that a relationship exists between flame emissivity and fuel properties. An explanation is provided for the finding that the carbon particle concentration on the flame axis of geometrically similar flames in larger size furnaces is smaller for identical fuels.—D. L. C. P.

REFRACTORY MATERIALS

New Basic Refractories Plant of Kaiser Aluminium and Chemical Corporation. (*Indust. Heating*, 1957, **24**, Feb., 371–379). A description is given of plant for manufacture of chrome-magnesite bricks at Columbiana, Ohio.—A. D. H.

Fundamental Reactions Occurring during the Manufacture and Use of Chrome-Magnesite Brick. R. G. Richards, A. Gunn, N. E. Dobbins. (*Trans. Brit. Ceram. Soc.*, 1956, **55**, Aug., 507–527). Numerous chrome ores with additions of 0 to 50% MgO were investigated.—D. L. C. P.

Developments in Refractories Research for the Iron and Steel Industry. (*Refractories J.*, 1957, **33**, June, 232–236). Topics from the BISRA Annual Report are presented.

Plastic Refractory in Slab Heating Furnaces. R. A. Smith. (*Iron Steel Eng.*, 1957, **34**, May, 99–102). The uses of plastic refractories in slab heating furnaces are discussed. The author claims greater furnace availability, harder furnace driving, greater output and longer life of the refractory.

Melting Relations of the Common Rock-Forming Oxides. J. F. Schairer. (*J. Amer. Ceram. Soc.*, 1957, **40**, July, 215–235). Diagrams for the systems Al_2O_3–SiO_2, MgO–SiO_2, CaO–SiO_2, FeO–SiO_2, CaO–Al_2O_3–SiO_2, MgO–Al_2O_3–SiO_2, CaO–MgO–SiO_2, CaO–FeO–SiO_2, MgO–FeO–SiO_2, FeO–Al_2O_3–SiO_2, and alkali metal oxide systems are given.

Properties of Refractories. (*Indust. Heating*, 1957, **24**, Feb., 359–360). Brief summaries are given of papers dealing with the formation of mullite, alumina–silica mortars, changes in length when castable refractories are heated and cooled, the cooling porperties of dozzle heads and the performance of ceramic tools.—A. D. H.

Determination of the Elastic Constants of Refractories by a Dynamic Method. J. R. Lakin. (*Trans. Brit. Ceram. Soc.*, 1957, **56**, Jan., 1–7). The method involves the determination of the resonant frequency of a regular specimen vibrating longitudinally, torsionally or flexurally.—D. L. C. P.

Determination of the Modulus of Elasticity of Refractories at High Temperatures. J. R. Lakin and C. S. West. (*Trans. Brit. Ceram. Soc.*, 1957, **56**, Jan., 8–13). An apparatus for measuring the modulus of elasticity of refractory specimens at high temperatures is described, and some results are given for silica and basic bricks. The significance of changes in modulus with temperature is discussed.—D. L. C. P.

The Relation Between Modulus of Elasticity and Abrasion Resistance in Refractory Materials. C. Storey and J. Mackenzie. (*J. Brit. Ceram. Soc.*, 1957, **56**, Jan., 14–17). It is suggested that a relationship exists between the modulus of elasticity and the abrasion resistance of refractory bricks.

The Mechanical Properties of Basic Refractories at High Temperatures. W. F. Ford and J. White. (*Trans. Brit. Ceram. Soc.*, 1957, **56**, June, 309–330). Thermal cycling of chrome–magnesite brick between 1150–1350° C (inducing oxidation-reduction of iron oxide) produces growth and friability and a large reduction in modulus of elasticity whereas this does not occur with magnesite. In bend at 1350°, chrome–magnesite showed greater creep resistance, but in tension, delayed fracture occurred even at 3·3 lb/in² within 12 hr at 1425°.

Loss in Strength of Silica Bricks Reheated in Hydrogen or Carbon Monoxide. K. Grant and W. O. Williamson. (*Trans. Brit. Ceram. Soc.*, 1957, **56**, June, 277–295). Mechanical and microscopical studies are reported on Australian bricks heated 4 hr at 1200°. Colour changes are also noted. Weakening may be due to calcium silicate crystallization.

Thermal Conductivity of Some Refractory Materials. J. F. Clements and J. Vyse. (*Trans. Brit. Ceram. Soc.*, 1957, **56**, June, 296–308). Measurements on silica bricks, carbon blocks, sillimanite, fused alumina, SiC and magnesite up to 1200° C and the effects of physical properties and other factors are reported.

CO Attack on Firebricks: The Physical State of the Deposited Carbon. W. R. Davis, R. J. Slawson and G. R. Rigby. (*Trans. Brit. Ceram. Soc.*, 1957, **56**, Feb. 67–85). Experiments on the nature of carbon deposited by CO on iron oxides are described. Three forms of thread-like carbon were identified, and compared with long-chain hydrocarbon polymers. At 450° C the catalyst is either cementite or iron percarbide.—D. L. C. P.

High Temperature Propane-Oxygen Furnace. (*Ceramics*, 1957, **9**, Feb. 14–16). The design of a high temperature furnace for firing refractory oxide samples up to 3600° F is given. The firing chamber, of stabilized zirconia, can be up to 9 × 6 × 8 in.—D. L. C. P.

IRON AND STEEL—GENERAL

Ironmaking in the U.S.S.R. E. W. Voice and S. Klementaski. (*J. Met.*, 1957, **9**, Apr., Section 1, 592–596). The authors give personal impressions of the Russian ironmaking industry, following the visits of two teams of British technicians to U.S.S.R. in 1955 and 1956. Present operating and design factors are outlined, and an indication is given of future development.—G. F.

Second All-State Conference of Steelworkers. V. Rybář. (*Hutník*, 1956, **6**, (11), 321–323. [In Czech]. A report is given of the topics discussed and the recommendations made to the industry at the Conference, which was held in Prague in the Autumn of 1956. Recommendations related mainly to (*a*) introduction of all-basic O.H. roofs, (*b*) increases in the alumina content of acid bricks, (*c*) increasing the rate of production of all-basic bricks, (*d*) researches on O.H. furnace design, (*e*) increasing the use of mechanized processes in O.H. repairs and reduction of repair times, (*f*) appointment of a committee of experts to review and advise on automatic control devices, (*g*) expansion of the utilization of stainless steel scrap in the O.H. furnace by the use of O_2, (*h*) accelerating the introduction of vacuum casting into foundries, (*i*) increasing the rate of production of pyrometers and their use for continuous bath temperature measurement and control, (*j*) investigating possibilities of more effective utilization of low grade blast-furnace sinters, (*k*) the introduction of further automation, remote control and measuring devices into steelworks, particularly utilizing telecommunication systems and television wherever appropriate, (*l*) increased production of mathematical computers for use in process control and in the works laboratory, and (*m*) increasing the quality of transformer sheet.—P. F.

The Work of the Iron and Steel Industry Must be Improved. (*Stal'*, 1957, (1), 3–4). [In Russian]. The achievements and failures of the Soviet Iron and steel industry in 1956 are outlined and measures for improved operation in 1957 are discussed. In 1956 the average value of the coefficient of utilization of useful blast-furnace volume was 0·78 (2·4% better than for the previous year), the production of steel per m². of O.H. furnace bottom area was 7·19 tonnes (3·3% better). Furnace idle-time figures were poor at many works and production costs did not decrease to the extent planned. Better supplies of ore and of high-basicity fluxed sinter are among measures recommended for blast-furnace practice in 1957, and for the industry as a whole more vacuum treatment of steel, better rolling procedures and more automation are proposed.—S. K.

Research at the Ukraine Institute of Metals. S. F. Kas'yanov. (*Stal'*, 1957, (6), 574–575). Work on the blast furnace, steel production, rolling mills, improvement in the transport of metals and on methods of investigation of the quality of metals is described.

Investigation at the Stalinsk Works. P. G. Glazkov. (*Stal'*, 1956, (9), 851–853). [In Russian]. A survey is given of recent investigations at an integrated iron and steel works in the South of the U.S.S.R. Blast-furnace, steelmaking and rolling practice have been improved through the joint efforts of plant and laboratory personnel.—S. K.

The Problems Before the Magnitogorsk Metallurgists. F. Voronov. (*Stal'*, 1957, (1), 5–6). [In Russian]. The director of the Magnitogorsk Metallurgical Combine surveys the achievements and failures at the combine in 1956 and discusses problems awaiting solution. The article is published in connection with the 25th anniversary of the combine's foundation.—S. K.

Creative Activity of Rationalizers and Inventors of the Magnitogorsk Metallurgical Combine. V. I. Golchin. (*Stal'*, 1957, (1), 93–94). [In Russian]. Some examples are given of inventions and improved operating methods proposed and adopted recently at Magnitogorsk. In the first six months of 1956, 2200 suggestions were made and 1300 adopted, leading to an estimated saving of 10^7 roubles.—S. K.

Report of the Activities of the Verein Deutscher Eisenhüttenleute During 1956. (*Stahl u. Eisen*, 1957, **77**, May 16, 609–642). A very detailed report on all activities of the Verein and the Max-Planck-Institut für Eisenforschung. (244 references).—T. G.

All-Union Conference on the Automation of Production Processes in Ferrous Metallurgy. A. K. Adabash'yan. (*Stal'*, 1956, (10), 942–945). [In Russian]. A conference was held at Magnitogorsk from 29th May to 4th June, 1956, which was attended by 562 delegates from Soviet industry and research and planning organizations and by delegates from other countries. A total of 96 reports were presented and discussed, 14 of them in plenary session. The sections at the conference included: blast-furnace, steel melting, rolling, tube-rolling, by-product coking, refractories, re-heating and heat-treatment furnaces, mining and steel-plant energetics. These discussed the theory and practice of automation for their fields. Reasons for insufficient existing automation were discussed and measures to be taken decided.—S. K.

Developments in Yugoslavia's Steel Industry. B. Djukic. (*Min. J.*, 1957, **248**, May 24th, 654–655). A short illustrated account of modern developments in the Yugoslavian steel industry is given. The Zenica plant is the largest and most modern, its output was 120,000 t. of iron and steel products in 1956.—B. G. B.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

Construction of the Blast Furnace 2000 cu. m. in Volume. D. S. Kashchenko. (*Stal'*, 1956, (10), 872–874). [In Russian]. It is pointed out that several of the constructions and operating methods in the recently made proposal for a 2000-cu. m. blast furnace are as yet untried, whilst others are unsuitable for the productivity and raw materials envisaged.—S. K.

All-Union Blast Furnace and Sinter Plant Conference, 1957. Ya. I. Dashevskii and N. K. Leonidov. (*Stal'*, 1957, (6), 501–502). Reports on the 1719 m^3 and the projected 2286 m^3 furnaces were presented, high temperature stoves, O_2 enrichment with high top pressure, new profiles, desulphurization, automatic control of gas flow, of scale cars and the use of dust-proof cabins were all discussed. Self-fluxing sinter was reviewed and the need to increase basicity until limestone was not needed in the blast furnace charge. Strong, easily reducible highly basic sinter from a variety of ores is required. New concentration plants should aim at $> 60\%$ Fe and $< 10\%$ SiO_2, the Fe should be raised to 63–65% and the SiO_2 lowered to 5–7%. Recommendations were, the addition of burnt lime to the sinter mix, reduction of lump size of feed, variation of fuel used with bed thickness, improved screening and transport of hot sinter dispensing with water quenching. Fluidized transport of fines, flue dust and fuel is to be developed. In the blast furnace, decreased slag, increased burden uniformity, exclusion of limestone and decreased coke ash and sulphur, increased blast temperature (over 1000° C), increased top pressure and development of automatic control of the gas stream.

Rational Distributor for Blast Furnace Burdens. N. S. Shchirenko. (*Stal'*, 1957, (6), 500–501). Uniform distribution is attained in the hopper of the small bell either by rotation during filling or by fitting a rotating intermediate hopper working during discharge from the skip.

New Design of Charge Distributor. A. S. Ayukov. (*Stal'*, 1956, (11), 975–976). [In Russian]. The design of McKee-type tops used in the U.S.S.R. is criticized and a top of new design is proposed. Model experiments with the proposed design showed good results and it is claimed that it would be easier to make gas-tight at high top pressures than existing tops.—S. K.

Automatic Loading and Weighing of Raw Materials for the Blast Furnace with Belt Conveyor Feed to the Skips. A. A. Ganich, V. F. Zarubin, and V. G. Yakovlev. (*Stal'*, 1957, (6), 496–500). Replacement of scale cars by weighing hoppers and a belt conveyor to the skips increases charging efficiency and reduces the amount of metal in the structure. The feeding systems for various sequences of ore and coke are shown. If cold sinter is charged instead of hot a further saving in structural weight is effected.

Charging Raw Materials to Blast Furnace Skips by Belt Conveyors. N. S. Fil' and I. S. Lemberikman. (*Stal'*, 1957, (6), 493–495). Replacement of scale cars by conveyors is being developed, and a diagram of the arrangement is given.

An Automatic Blast Furnace Charging Control System. (*Iron Steel Eng.*, 1957, **34**, May, 149–153). The article describes a new automatic blast furnace charging control developed by A. G. McKee and Co. The system consists essentially of 3 panel boards, which include a scale car operator's panel, a manual and transfer panel and a hoist panel.

Graphical-Analytical Investigation of the Tipping Curves of Blast-Furnace Hoists. Ya. F. Chel'tsov and G. A. Dubrovin. (*Stal'*, 1956, (9), 777–779). [In Russian]. A theoretical analysis of the working of a new design of skip-tipping arrangement used at the Zaporozhstal' works is presented. Its use at other works is recommended.—S. K.

The Importance of Measuring and Weighing in Determining the Composition of Blast-Furnace Burdens. V. Friedrich. (*Hutník*, 1956, **6**, (11), 329–334). [In Czech]. The principles leading to the correct composition of the charge, assuming given raw materials, are discussed, and the potential losses in thermal and economic efficiency arising out of faulty charge composition are considered.—P. F.

Service Life of Blast Furnaces. N. G. Molchanov. (*Stal'*, 1956, (11), 972–974). [In Russian]. This article continues discussion on refractories and cooling systems for the blast furnace, and deals critically with proposals made by S. I. Shkumatov in *Stal'*, 1956, (3), 200–203. The present author recommends high-alumina or carbon linings for the hearth and hearth pad and high-alumina (iron-free) linings with an outer layer of iron-free thermal insulating material for the stack, and claims that this would enable stack cooling to be dispensed with. If stack coolers are used they should be of the peripheral plate type as proposed by Oreshkin (*Metallurg*, 1956, (3)), since conventional stack coolers encourage refractory failure by creating zones favourable to carbon deposition and resulting in temperature gradients conducive to cracking.

Possibilities for the Use of Calculating Machines in the Automatic Control of Blast Furnace Performance. I. A. Rylov. Although automatic control has been developed considerably, no single parameter has been found to characterize the working of the furnace. One method is to divide up the operation into unit processes and base control on pressure drop in stack and control of distributing gear etc. The other is to use calculators. An equation is used for thermal and materials balances and from it are calculated output per min, and coke consumption per t. iron and ore. The desirability of a coding system is outlined and a programme for further development is indicated.

Automation in the Iron and Steel Industry. R. Perlick (*Neue Hütte*, 1957, **2**, Feb.-Mar., 92–97). Problems of blast furnace control are adumbrated. Proposed solutions are discussed. Automatic control has been successfully applied to O.H. furnaces. The use of commercial oxygen necessitates the installation of control and recording instruments in converters. Research remains to be done on automatic control of electric furnaces. The automation of rolling mills depends on technical developments which in turn depend on economic factors.—L. J. L.

Automatic Regulation of Blast-Furnace Working. G. G. Piotrkovskiĭ and F. V. Ashikhmin. (*Stal'*, 1957, (1), 16–17). [In Russian]. The authors describe the scope and methods of automation introduced in the blast furnace plant at Magnitogorsk. For clean-gas pressure regulation standard hydraulic regulators operating a valve are used. Stove temperature is regulated by the e.m.f. of dome-temperature thermocouples, an electromechanical potentiometer, distributing mechanism and actuating mechanisms operating a valve in the gas inlet line. For hot-blast temperature the standard Soviet arrangement is used. In the blast-moisture regulating system the transducer-psychrometer is of the spray type with stabilization of sample-air pressure and temperature and of tank water-level. The top-pressure is controlled through electrical regulators working d.c. actuating mechanisms. A blockage-proof system for protecting the big bell operation when back-pressure is absent is used.—S. K.

The Theoretical Limitations of Blast Furnace Operation. H. Suematsu. (*Tetsu to Hagane*, 1957, **43**, Jan., 3–11). [In Japanese]. The results are given of tests made with present-day raw materials. With assumed values for ash in coke, mean Fe value of the ore, blast temperature and utilization of top gas, values are given for the limits of blast-furnace operation in terms of output, coke rate and wind blown. For the most efficient reduction, the ratios of direct to indirect reduction, and of CO to CO_2 in the top gas, are calculated. The figures given can vary with coke quality, ore quality, and operating methods.—K. E. J.

On the Movement of the Charge Column in the Blast-Furnace. V. K. Gruzinov. (*Stal'*, 1956, (9), 771–773). [In Russian]. The following are the main conclusions drawn from a critical evaluation of the influence of various factors on the descent of the charge in the furnace: almost half (and sometimes more) of the movement of the charge in modern practice is on account of fusion; relations obtained by G. M. Malakhov can be used to describe the movement above the combustion zones; friction against the walls depends, in the range of lines used, more on stack than on bosh design; static pressure determinations at the tuyere level either in models or full-scale cannot be used for deciding the suitability of the lower parts of the furnace, but the active-weight coefficient fully characterizes the stack lines; the role of the bosh in hindering the descent of materials has been greatly exaggerated.—S. K.

A Study of the Tuyere Combustion Zone. J. Taylor, G. Lonie, and R. Hay. (*J. Iron Steel Inst.*, 1957, **187**, Dec., 330–341). [This issue].

Some Observations on the Operation of Blast Furnaces with and Without Tuyere Blocks. G. von Struve (*Neue Hütte*, 1957, **2**, (Feb.-Mar., 138–142). The disadvantages of hori-

zontal cooling elements are enumerated. External water spraying, and thin-walled linings are advocated.—L. J. L.

New Investigations of the Coke Combustion Process at the Blast-Furnace Tuyeres. M. Ya Ostroukhov. and L. Z. Khodak. (*Stal'*, 1956, (10), 867–872). [In Russian]. Recent work, much of it non-Russian, on the combustion and movement of coke in front of the blast-furnace tuyeres is reviewed and some factors influencing the extent of the oxidizing zone are discussed.—S. K.

Effect of Sinter on Blast-Furnace Performance at Margam Works of the Steel Company of Wales Ltd. J. S. McLean. (*Iron Coal Trades Rev.*, 1957, **174,** June 14, 1375–1379). Comparative furnace performances are shown and the effect of sinter on operating conditions is described.

Investigation of Heat Transfer and Reduction Processes when Operating on Fluxed Sinter. N. N. Babarykin and F. A. Yushin. (*Stal'*, 1957, (1), 7–15). [In Russian]. In this article the methods and results are presented of extensive blast-furnace exploration work carried out at Magnitogorsk in 1955, in connexion with the use of hot sinter of increased basicity. The work included materials and gas-sampling and temperature measurement at various levels in three furnaces. It was found that the use of this material in place of limestone, Mn ore and O.H. slag led to increases in charge and gas temperature near the top of the furnace. When raw limestone was present in appreciable quantities the temperature in the part of the stack acting as a heat-transfer reserve approximated to that of its decomposition; after complete elimination of the stone the temperature of this reserve height is fixed by the minimal value for rapid reaction of CO_2 with coke carbon and rises to 1100–1150° C. In relatively burden-rich regions stationary CO_2 concentrations were found so close to these in equilibrium with FeO/Fe as almost to stop reduction; to eliminate this effect screening of all fines from the ore and sinter and better distribution is recommended. Average degrees of reduction in the magnetic fraction of samples taken from levels 4, 10, 16, and 19 m. below the big bell were 20·5, 29·4, 57·0, and 84·7%, respectively, the proportion of the respective furnace cross-sections sampled being 95·6–98·0%. Heat transfer rate was found to follow a similar quick-slow-quick pattern. It was found that as the charge descends, the burden materials break up and separate into iron-rich and poor regions, this effect facilitating primary slag formation, but reducing gas permeability. It is considered that gas-entrained slag from lower levels plays an important part in primary slag formation.—S. K.

Blast Furnace Operation with a High Top Pressure of 0·8 atm. (gauge). P. F. Sergeev and N. I. Vasil'chenko. (*Stal'*, 1956, (11), 963–971). [In Russian]. The top pressure on nos. 1 and 2 blast furnces at Novo-Lipetsk was increased from 0·6 to 0·8 atm. gauge towards the end of 1954. This article describes the effects of this change and the modifications of ancilliary equipment carried out. The furnaces smelt a burden with about 50% Fe and containing 50–52% of 0–10 mm. fines. The change enabled the productivity of the furnaces to be increased by 3·8–5·6%, blast temperature to be raised by 32–63° C and the coke rate and flue dust to be decreased by 2·3–1·3% and 7·2–19·5%, respectively.

Thermal Effects of Steam Additions to the Blast on the Processes Occurring in the Hearth Zone of the Blast-Furnace. M. Prouza. (*Hutnické Listy*, 1956, **11,** (10), 577–579). [In Czech]. The use of artificially moistened blast is considered on the basis of existing experience and from the point of view of thermodynamics. It is shown that strict control of humidity is required, a high steam content of the blast necessitating a high blast temperature. Otherwise a deterioration in the coke-rate may result.—P. F.

Cooling of Blast-Furnace Agglomerates. H. Krumnikl. (*Hutník*, 1956, **6,** (10), 289–296). [In Czech]. The problems of storage and transportation of agglomerates are surveyed, and various possibilities of cooling are reviewed from the technological and economic points of view. The choice of method of cooling is shown to be determined to a large extent by local conditions and by the properties of the agglomerates used. The author recommends extensive laboratory tests on the sintering behaviour, from which the most appropriate cooling methods may become apparent, e.g. cooling rates may be determined which would not lead to undue crumbling of the agglomerates due to thermal shock.—P. F.

PROPERTIES, TREATMENT AND USE OF SLAGS

Further Considerations on High Alumina Blast Furnace Slags. H. Schrader and P. Mehta. (*Trans. Indian Inst. Met.*, 1954–55, **8,** 61–79). Increased beneficiation treatment of ores has led to increase in the alumina content of slags and the use of high or low basicities to maintain fluidity with slags containing 35% Al_2O_3 is discussed. Fluidity measurements conflicted with conclusions drawn from liquidus temperatures for the $CaO–MgO–SiO_2–Al_2O_3$ system. The addition of up to 4% TiO_2 increased fluidity at 1400°–1500° C. The fluidity of high-alumina slags could be correlated with microstructure. A fine needle-like structure was obtained in fluid slags suitable for operating conditions. Preliminary experiments showed that for high alumina practice, silicon reduction approaches equilibrium in practical furnace conditions but sulphur removal is far from equilibrium.—A. D. H.

Slag Analysis. S. Velken. (*Tidsskr. Kjemi Bergvesen Met.*, 1957, **17,** No. 2, 20–23). [In Norwegian]. A brief survey is given of the literature dealing with the analysis of slags obtained in the manufacture of iron, steel and ferroalloys.

Recovery of Manganese from Steelmaking and Blast-Furnace Ferromanganese Slags. N. J. Wadia, S. Visvanathan, and V. G. Paranjpe. (*Trans. Indian Inst. Met.*, 1954–55, **8,** 81–108; discussion 108). The recovery of Mn from blast furnace slags by reduction with C, Si or Al was investigated and products like ferro-manganese, silico-manganese, and 'Simanal' (low in iron) were obtained. High sulphur-ferromanganese slags were roasted at 650° C to remove 10–15% Mn as water-soluble $MnSO_4$ suitable for the production of electrolytic Mn. Acid O.H. slag can be treated by preferential reduction of iron followed by smelting of the manganese-rich-slag to produce ferromanganese. Basic O.H. slags require production of spiegeleisen, followed by preferential oxidation of Mn and the oxidized slag is then reduced to ferromanganese.—A. D. H.

Production of Portland Cement from Slag Produced during the Smelting of Low Grade Iron Ore. (*Canad. Mines Techn. Surveys Research and Special Projects*, (July-Dec. 1956), 1957, Feb. 8, 18–19).

DIRECT PROCESSES

On the Reduction of Iron Ores with Gaseous Reduction Agents. M. Omar, W. Machu, and S. Y. Ezz. (*Arch. Eisenhüttenwesen*, 1957, **28,** Mar., 117–122). The reduction of Egyptian iron ores from Asswan and Wady-Carim with H_2 waste gases and reformed cracking gases from petroleum refineries has been studied. A tubular furnace was used for the tests. The ores were more readily reducible with H_2 than with the cracking gas. Uncracked gas from the refineries is not suitable for reduction. The rate increases as the temperature increases, particle size decreases, and gas velocity increases.—L. J. L.

The H-Iron Process. A. M. Squires and C. A. Johnson. (*J. Met.*, 1957, **9,** Apr., Section 1, 586–590). A description is given of the pilot plant operated jointly by Bethlehem Steel Co. and Hydrocarbon Research Inc. for directly reducing iron ore fines with hydrogen (the "H-iron" process). The principles of the process and operating experiences are discussed, and details are given of the advantages and economics of the process.—G. F.

PRODUCTION OF STEEL

Steel from the Black Country. From Wrought Iron to Tube Billets. (*Engineering*, 1957, **184,** July 19, 86–87). An account of the Round Oak Steelworks, Brierley Hill.

The International Development of Instruments and Controls for the Iron and Steel Industry. H. Krüger. (*Neue Hütte*, 1957, **2,** Feb.-Mar., 103–111). A selection of instruments for measurement and control are described and discussed. A number of recently built plants is described. Instrumentation and control of blast furnaces, O.H. furnaces, and 3-zone gravity-discharge furnaces are explained in detail. The use of analogue computers and of punched cards in the control of rolling mills are described.—L. J. L.

Dependence of the Alumino-thermic Process Temperature on Specific Heat. V. A. Bogolyubov. (*Stal'*, 1957, (6), 531–535). Using a W–Mo thermocouple the temperature of the thermite reactions was measured and the heat evolution

by means of a calorimeter and a straight-line relation was confirmed. This relationship can be used to make a rapid estimate of the temperature.

Human Aspects of Open-Hearth Production. G. Grosvenor. (*Amer. Iron Steel Inst. Reg. Tech. Meetings*, 1956, 361–367). An account of labour relations in relation to production at the Colorado Fuel and Iron Corp.

Basic Checkering from the Thermal Standpoint. H. Zahorka. (*Radex-Rundschau*, 1956, (6), 272–285). A method of calculating regenerator heating capacity is demonstrated for the case of a 70-ton O.H. furnace. It is concluded that this furnace has insufficient regenerator capacity but that the use of basic checkering would increase the checker heating capacity substantially.

The Application of Metallic Recuperators to Industrial Furnaces. W. Ernest and C. B. Tupholme. (*West Scotland Iron Steel Inst.*, Paper No. 500, Jan. 11, 1957). The authors consider the theoretical and practical aspects of design with particular reference to heat transfer coefficients, parallel and counter flow, critical partition temperatures, pressure drop and recuperator efficiency. The constructional materials now available (18/8 Ti stabilized steel, 25/16 steel, 25/20 steel, 20 Cr steel with or without titanium) are discussed and their applications and limitations are indicated. Typical commercial designs are described and an account is given of operating experience of a recent installation. The authors conclude with some suggestions for consideration in future installations (11 references).—L. E. W.

New Information on the Use of Manganese in Steelmaking. J. Kašpar. (*Problems and Perspectives of Czechoslovak Metallurgy and Foundry*, 1956, 219–230). The effect of Mn on desulphurization is discussed though further investigation is needed. Under normal conditions pig iron even with low-Mn in the charge gives good results but if more MnO is needed in the charge some manganese ore is added, especially where the blast furnace burden is low in Mn.

Equipment for the Mechanical Stirring of Molten Metals. P. Wincierz. (*V.d.I.Z.*, 1956, **98**, Dec. 1, 1881–1882). Mechanical oscillation during melting helps the outgassing and purifying of the metal. The method by which mechanical stirring is obtained by subjecting the melt to electromagnetic forces is described. Even with relatively simple equipment it is easy to obtain 50% efficiency of energy transfer.—J. R. P.

Recovery of Manganese from Open Hearth Slag by Chloridization Processes. B. P. Schofield. (*Amer. Iron Steel Inst. Reg. Tech. Meetings*, 1956, 231–268). The HCl and $CaCl_2$ processes are outlined, the latter through the pilot-plant stage to commercial possibility.

Alloy Steel Production by Combination of Open Hearth and Electric Furnaces. F. Pobořil. (*Problems and Perpectives of Czechoslovak Metallurgy and Foundry*, 1956, 212–218). The origins of the process are described and its application to the production of special qualities of steel for which the O.H. is unsuitable and the capacity of electric furnaces insufficient. It should be suitable for production of transformer and alloy steels for large ingots where uniformity, perfect deoxidation and high purity are required.

Experimental Electric Smelting of Ores and Related Materials at the Department of Mines and Technical Surveys, Ottawa. G. E. Viens, R. A. Campbell and R. R. Rogers. (*Canad. Min. Met. Bull.*, 1957, **50**, Feb., 70–77). A 3-phase, 250 kv.a furnace has been used for quantitative smelting studies on iron, manganese-iron, titanium-iron, nickel and chromium-iron ores, also industrial by-products. The results of the trials are set out and findings summarized.

Protection of an Electrical Resistance Furnace. L. I. Sapiro. (*Stal'*, 1956, (6), 569–570). [In Russian]. Defects likely to develop in industrial (several hundred kW) resistance furnaces are listed and measures for avoiding them briefly considered.—S. K.

Source Life of Acid Linings in High-Frequency Furnaces. M. Wald. (*Problems and Perspectives of Czechoslovak Metallurgy and Foundry*, 1956, 231–238). Life is related to steel composition and a method for evaluating the lining and measuring porosity has been developed.

Metallic Materials. E. Siebel. (*V.d.I.Z.*, 1956, **98**, Sept. 21, 1617–1621). A review of the latest developments in alloys, especially steel (constructional, mild steel, tool steel, heat resisting steels and special steels). Cast and sintered iron are also included as well as the non-ferrous materials replacing aluminium in the construction of supersonic aircraft. (91 references).—J. R. P.

Low-Carbon Bainitic Steels. K. J. Irvine and F. B. Pickering. (*J. Iron Steel Inst.*, 1957, **187**, Dec , 292–309). [This issue].

Special Steels. L. Dainelli. (*Rivista Mecc.*, 1954, **5**, Oct. 23, 7–16). [In Italian]. The author reviews the various special steels used for cutting tools. These are classified by type according to their uses and whether for hot or cold working. Considerable technical data are given.—M. D. J. B.

Low-Alloy Structural Steels MK and M. V. G. D'yakov and S. I. Vol'fson. (*Stal'*, 1956, (7), 655–656). [In Russian]. The mechanical properties of sheets and seamless tubes of MK (0·10–0·15% C, 1·2–1·4% Mn, 0·6–0·8% Si, 0·10–0·15% Cr, 0·2–0·3% Ni, 0·5% Cu, 0·010–0·015% S and 0·020% P) and M (0·10–0·11% C, 1·39–1·47% Mn, 0·17–0·19% Si, 0·18% Ti, 0·09–0·13% Cr, 0·13–0·23% Ni, 0·27–0·31% Cu, 0·020–0·021% S and 0·024–0·035% P) low-alloy steels have been studied.—S. K.

Technical and Economic Aspects of the Use of Stainless Steels. C. M. Lerici. (*Rivista Mecc.*, 1955, **6**, Sept. 3, 14–17). [In Italian]. The author discusses the various points to be taken into consideration when choosing stainless steels. Cost and their availability are examined. Future trends and possible developments are reviewed.—M. D. J. B.

West Tests Low Nickel 200 Series Stainless. K. Elers and T. A. Woolsey. (*Western Metals*, 1957, **15**, Jan., 23–26). Compositions, properties, corrosion-resistance, fabrication methods and welding of Cr–Ni–Mn steels are described.

A New Nickel-Free Stainless. E. J. Whittenberger, R. H. Witchey and R. C. Rudolphy. (*Product Eng.*, 1956, **27**, Dec., 151–153). An account of Tenelon, an austenitic steel is described and its mechanical properties and corrosion-resistance given.

Nickel Economy in Melting Stainless Steel. E. I. Astrov. (*Stal'*, 1956, (7), 651). [In Russian]. In this brief note the chromium/nickel ratio is shown to have little effect on ferrite formation in type 1Kh18N9T stainless steels. Recent views on nickel economy are critically described and difficulties of composition control are considered.—S. K.

Steels with Very High Elastic Limits. (*Aciers Fins Spec. Franç.*, 1956, June, 20–22). Developments in the aircraft industry have led to the increasing use of steels with a high elastic limit, and details are given of the composition and characteristics of a steel developed to meet this requirement.

Super-Strength Steels. W. B. Kemmish. (*Machinery*, 1956, **89**, Dec. 21, 1421–1422). A brief review is given with a general account of the composition and heat-treatment of newer alloy steels, particularly one with 0·4% C, 1·3% Mn, 2·0% Si, 1·4% Cr, 0·2 V and 0·35% Mo said to have a tensile strength of 130 t/in².

Alloying Steels 18 KhGT and 30KhGT with Zirconium. G. L. Livshits and G. A. Torpanova. (*Stal'*, 1956, (6), 571). [In Russian]. The small quantities (0·05–0·08%) of zirconium, through its inhibition of grain-growth, enabled higher hot-working and carburizing temperatures to be used for steels containing about 1·1% Cr and 0·08–0·15% Ti.—S. K.

The Use and Misuse of Moulds. J. Thomson. (*West Scotland Iron Steel Inst.*, Paper No. 497, Dec. 21, 1956). Factors in the design of ingot moulds are discussed, including some special designs, materials and the composition of the metal to be cast. The use of the mould in the casting pit is considered and the types of failure described. Details are given of average life of moulds of sizes from 97 to 26 in. diameter.—L. E. W.

Determination of Casting Speed in Pouring Ingots from the Top Across an Intermediate Ladle. A. Děkanovsky. (*Problems and Perspectives of Czechoslovak Metallurgy and Foundry*, 1956, 251–262). Information on viscosity is given, but there are also other factors concerned, and for any type of steel there is a method for any type of steel there is a method and note of casting and a diameter for the main and intermediate ladle spouts giving the best results which need to be practically determined and checked against the theoretical values.

Studies on the Ingot Making Process in Nitrogen Atmosphere. I. J. Takao, T. Shimose and K. Saito. (*Tetsu to Hagane*, 1956, **42**, Oct., 956–961). [In Japanese]. Nitrogen blowing is effective in preventing oxidation of molten steel during pouring, particularly as compared with floating boards. Some oxidation occurs in the mould, as shown by scum formation. (22 references).—K. E. J.

The Development of the Process of Continuous Casting of Steel. M. S. Boichenko, V. S. Rufes and V. V. Fulmakht.

(*Metallurg*, 1956, (2), 7–11). [In Russian]. Members of the Central Scientific Research Institute of Ferrous Metallurgy in cooperation with two metallurgical works have carried out experiments of continuous casting of steel. In the Novo-Tula works the underground installation was designed for the production of steel slabs 150 × 500 mm in cross section, with a velocity of 0·5 to 1·5 m/min and an output of 16 to 53 tons/hour. The casting was done from a ladle of 10 tons capacity. The primary cooling was done in a water-cooled mould from which the steel band passed, supported on the sides by rollers, into a secondary open air cooler containing water atomizers. This method of cooling prevented the formation of internal cracks. Between levels 8 m and 14 m the steel band was cut by oxy-acetylene cutters and at a lower level the slabs were raised by a lift. A different plant was designed at the Krasnoe Sormovo works, where the casting was simultaneously produced in two bands, 175 mm × 420 mm in section and 1920 mm in length cast from a 50 tons ladle and a total capacity 45–55 tons/hour. The lifts were arranged in a V fashion and the work was automatically controlled from remote control panels. The internal structure of the metal during casting was studied by means of radioactive isotopes of phosphorus and sulphur.—S. I. T.

Hydrodynamics of Liquid Steel in the Ingot Mould. A. A. Zborovskiĭ, L. K. Strelkov, M. K. Skul'skiĭ, and E. I. Rabinovich. (*Stal'*, 1957, (1), 24–30). [In Russian]. The object of the investigation described was to find the duration and study the character of natural convection in the mould during the solidification of 7-ton ingots. ^{59}Fe and ^{32}P were used as tracers. It was found that with killed steel the colder layers at the solidification front descend and displace hotter metal along the ingot axial region; with rimming steel peripheral metal ascends and descends along the axial region. Metal circulation near the top of the ingot continues for most of the freezing time. It is suggested that ingot-solidification calculations must take liquid-metal convection into account as well as conductivity.—S. K.

The Shape of the Shrinkage Head. V. G. German. (*Liteĭnoe Proizvodstvo*, 1956, (3), 32–33). [In Russian]. From a mathematical analysis of the freezing of metal in containers of various shapes it is concluded that spherical heads are fundamentally less suitable than flat ones.—S. K.

Geometry of the Bottom Part of an Ingot. L. V. Andreyuk. (*Stal'*, 1957, (1), 35–38). [In Russian]. Cropping losses in ingot rolling have been greatly reduced by modifications in ingot bottom configuration produced by using stools with spherical depressions. It is recommended that ingots of different shapes should be used for rolling wide slabs and squares.—S. K.

Rational Method for Breaking Away the Jet in Sheet Ingots. N. I. Vinogradov. (*Stal'*, 1956, (9), 847–849). [In Russian]. Crack formation in bottom-poured ingots with shoulders has been investigated. The use of shoulders for facilitating the break with the jet and ingot removal was found to be more suitable for section rolling than for sheet rolling where it leads to increased metal loss. For sheet rolling ingots only partial shoulders should be used, which are sufficient to secure the break-away but do not introduce the disadvantages of normal shoulders. For avoiding cracks below the shoulder the hot tops should be placed to leave the least possible gap, and pouring rate and metal temperature should be carefully controlled.—S. K.

Simulation Technique. (*Castings*, 1957, **3**, May, 19–20). Experiments to study steel casting on scaled-down castings of that or other materials are briefly described. Castings of a stearin–naphthalene mixture charged with CO_2 are shown.

Influence of Technological Factors in the Melting and Teeming of 1-2X13 Steel on the Presence of Hair Cracks. M. I. Vinograd, G. S. Chernyak, and N. D. Orekhov. (*Stal'*, 1957, (6), 560–562). Relatively hot metal (1550–1650° C) bottom poured gives a considerable reduction in hair cracks when only the upper parts of the ingots are used.

Some Technical Problems Relating to Continuous Casting of Steel. L. Šajch. (*Hutnické Listy*, 1956, **11**, (12), 705–709). [In Czech]. The importance of pouring temperature, billet withdrawal rates, methods of cooling etc. are discussed, and Soviet methods of casting steel by continuous methods are reviewed. Economic aspects of continuous casting and its advantages over conventional methods of bar production are considered.—P. F.

Solidification and Heat Removal Processes in Continuous Casting. A. A. Skvortsov, A. D. Akimenko, and K. P. Korotkov. (*Stal'*, 1956, (10), 883–890). [In Russian]. The investigation described on billet solidification and cooling in continuous casting was carried out in connection with the adoption of the process at the " Krasnoe Sormovo " works. Theoretical calculations on the process are complicated by the fact that heat loss from the billet surface varies in different parts of the continuous casting installation. A hydraulic integrator was used for studying solidification in the liquidus-solidus temperature range which enabled the course of thermal processes occurring during the solidification and cooling o flat billets to be studied. Results agreed well with exper mental data for surface temperature, thickness of solidified layer and other factors. By using this integrator the optimal parameters (e.g. rate of withdrawal of the billet) can be determined with sufficient accuracy to be useful for designing new installations. The overall heat transfer coefficient for secondary jet cooling with a water velocity of 4–5 m/sec is 3000 to 7000 cal/(sq. m. hr. ° C). For a billet-surface temperature range of 1000–600° C the coefficient rises appreciably because of intense evaporation, jet impact prevents a steam film forming. Decrease in the jet velocity and increase in surface temperature lead to a considerable lowering of the coefficient, which is particularly important if circulating cooling systems are contemplated. With " soft " jets steam-film formation lowers the coefficient. Further improvements are planned for the integrator.—S. K.

Research on the Non-Metallic Inclusions of Killed Steel. III. Y. Koike, I. Inuzuka, and T. Horigome. (*Tetsu to Hagane*, 1957, **43**, Feb., 122–126). [In Japanese]. The use of an experimental ingot mould with a refractory-lined bottom led to a reduction in streak seams at the bottom of the ingot. With varying Al additions, Fe–Mn–Al silicate seams, which had been molten in the mould, and Al_2O_3 seams, which had been solid, were found in the ingot. In a 5-ton ingot, seams hardly ever occurred in the columnar crystals, but were found at random within the ends of columnar crystals at the ingot bottom.—K. E. J.

Anisotropy of Mechanical Properties of a Steel Ingot Vacuum Melted and Cast. Ya. B. Gurevich. (*Stal'*, 1956, (9), 815–817). [In Russian]. The aim of the investigation described was to find the relation between the mechanical properties and the direction along which test pieces were cut out from the columnar zone of small ingots of 0·3 and 0·45% C and an alloy austenitic (type 1 Kh 18 N 9) steel. The melting and casting of the steels was carried out partly in vacuum and partly in a He atmosphere. Tension and torsion tests were made on specimens in the cast state and after hot forging to 70–75% reduction followed by heat treatment. Results are discussed in terms of inter- and intragranular effects.—S. K.

PRODUCTION OF FERRO-ALLOYS

Analysis of the Indices and Special Features of the Electric Melting of Ferrosilicon and Silichrome at Two Works. V. V. Rukavishnikov and V. A. Kravchenko. (*Stal'*, 1956, (11), 989–993). [In Russian]. Analysis of materials and electricity consumptions at the Chelyabinsk and Zaporozhe ferro-alloy works shows that in the period 1950–54 the practice at the former deteriorated more and more compared with that at the latter. This deterioration is attributed to insufficient immersion of electrodes and poor slag removal. Experience at Zaporozhe shows that elevated ferrosilicon slag viscosity can not be attributed to the presence of pieces of carborundum. The destruction of the middle and upper parts of the furnace stacks at Chelyabinsk is considered to be a consequence not of washing away of carborundum previously solidified on the walls, but of insufficient electrode immersion. An editorial note suggests that the explanations offered are incomplete.—S. K.

Heat-resistant Ferrosilicon with Nodular Graphite. A. Tiefenbach. (*Giessereitechn.*, 1957, **3**, Mar., 70). Nodular cast iron with 2·4–3·0% C and 6·0–7·0% Si having a ferrosilicon structure has good mechanical characteristics (tensile strength about 40 Kg per sq. mm), is highly resistant to scaling, and is stable up to 900° C.—L. J. L.

FOUNDRY PRACTICE

Casting Vessel with a Floating Slag Trap. P. V. Shtonda. (*Liteĭnoe Proizvodstvo*, 1956, (4), 26–27). [In Russian]. The

design of slag trap described ensures that slag in the first portions of metal entering the mould system is effectively freed from slag. Its adoption at a works has reduced rejects from 5·5–6·5% in 1954 to 2·7% in October, 1955.—S. K.

Dephosphorization of Cast Iron in Hydrogen. B. Chatterjee and P. C. Ghosh. (*Trans. Indian Inst. Met.*, 1954–55, **8**, 161–164). Removal of phosphorus from three samples of cast iron in bulk and powdered form was attempted by treatment in moist H_2 for 8 hr. at 1000° C. Dephosphorization was achieved only with one sample which indicates that the composition of the iron is of greater importance than physical or chemical factors.—A. D. H.

Processes for Working Metals by Pressure in the Liquid and Plastic State and Prospects for their Development. V. M. Plyatskiĭ. (*Liteĭnoe Proizvodstvo*, 1956, (6), 7–11). [In Russian]. The various pressure casting and forming methods are discussed from the aspects of product quality, surface finish and metal economy. The prospects of their further development are considered.—S. K.

Notes on the Planning of Foundries. I. B. Livshits. (*Liteĭnoe Proizvodstvo*, 1956, (3), 11–12). [In Russian]. Experience in the operation of large modern foundries is analysed as a guide to improved planning. Subjects briefly dealt with include the replacement of old equipment, material transport, hopper and bunkers, mechanization.—S. K.

Scandinavian Foundry Review Husqvarna. (*Foundry Trade J.*, 1957, **103**, July 18, 71–77). **Karlstads** (77–79). **A. B. W. Dan Bergman** (79–82).

Scandinavian Foundry Review. ASEA. (*Foundry Trade J.*, 1957, **103**, July 11, 41–50). An illustrated account, with plan, of the ASEA and Domnarfvet foundries in Sweden is given.

Ways of Increasing the Capacity of Iron Foundries. V. I. Sverdlov. (*Liteĭnoe Proizvodstvo*, 1956, (4), 29). [In Russian]. Iron foundry practice in the U.S.S.R. is briefly considered and the technical and economic effects of organizational and lay-out changes are discussed. A foundry with a total area of 5860 sq. m. is outlined.—S. K.

Cupola with a Slag Separator. M. G. Bezdol'nyi. (*Liteĭnoe Proizvodsvo*, 1956, (2), 25–26). [In Russian]. The use of a continuously operating slag trap has improved cupola practice. Both metal and slag leave the cupola by the iron notch and after separating by gravity in the trap, flow down different runners.—S. K.

Increased Efficiency in Iron Foundries Through the Use of the Hot-Blast Cupola. H. Kalpers. (*Maschinenwelt Elektrotechnik*, 1957, **12**, Jan., 1–3). The advantages of the hot-blast cupola are outlined. The importance of recuperator design is emphasized, and some industrial applications of the cupola are described.—L. D. H.

Induction Melting Puts Fabricator Into Gray Iron Production. H. B. Osborn. (*Mod. Castings*, 1956, **30**, Dec., 51). Improved quality was obtained by induction furnace installations.

Principal Types of Cast Iron and Broad Spheres of Application. J. G. Pearce. (*Brit. C.I. Res. Assoc. J. Res. Dev.*, 1957, **6**, Apr., 521–529). Flake, nodular, and special irons are reviewed and the classification considered in relation to B.S. specifications for grey castings, whiteheart and blackheart malleable and nodular castings.

Materials Engineering File Facts. Nodular or Ductile Irons. (*Mat. Methods*, 1956, **44**, Sept., 143–145). Properties of 6 grades are tabulated, with heat treatment, fabrication properties and uses.

Quality Cast Iron and Methods of Production. G. Giacomelli. (*Fonderia Ital.*, 1957, **6**, April, 149–151). [In Italian]. Cast irons are divided into grey irons; high-resistance cast irons including quality grey irons and alloyed irons; and malleable irons. The properties of the irons constituting these main groups are discussed and production methods are described.—M. D. J. B.

The Influence of Sulphur on the Degree of Saturation of Cast Iron and the Practical Use of the Degree of Saturation in the Operation of a Foundry. P. Tobias and H.-W. Wenig. (*Giesserei*, 1957, **44**, Feb. 14, 97–100). The development of methods for assessing the quality of cast iron is described from the introduction of the sum (C + Si) to the use of the degree of saturation. Several experiments are then described in which the effect of various percentages of sulphur on the quality of cast iron was investigated. The S_e-value is defined and its application described.—R. J. W.

What Iron Foundries Are Doing With the Injection Process. G. P. Dahm. (*Mod. Castings*, 1956, **29**, June, 50–53). Desulphurization, increase of C and introduction of inoculants for nodular iron production are discussed. Structures are shown and compositions given.

Injection Processes in American Foundries. G. P. Dahm and C. E. Bieniosek. (*Brit. Found.*, 1957, **50**, Apr., 190–196). Desulphurization and carburizing are reviewed in some detail.

Calcium Silicide and its Significance as an Inoculant For Grey Iron Castings. B. Körös. (*Kohászati Lapok*, 1954, **9**, Sept., 212–216). After a review of the manufacture of calcium silicide the causes of its favourable effects on inoculation, e.g. desulphurizing, degassing and deoxidizing as well as its superiority over ferrosilicon with high silicon content are discussed, and comment is made on the promising results obtained from the examination of inoculated grey iron piston rings, brake shoes and ingot moulds in Hungary.—P. K.

Research on the Treatment with Magnesium Nitride for Spheroidising Graphite in Iron. II. On the Preparation of Magnesium Nitride and its Properties. R. Yoda. (*Nippon Kinzoku Gakkai-Si*, 1954, **18**, Nov., 573–578). [In Japanese]. Conditions for producing magnesium nitride from its elements were explored; the best results were given by passing nitrogen at a given flow rate over magnesium turnings in an iron boat at 900° C for 1 hour. The behaviour of the compound on heating in air and *in vacuo* is reported. To produce prompt spheroidization, nitrides must be of high purity, prepared at above 800° C, and about 50 mesh in size. (22 references).

Some Experiments on Methods of Production of Nodular Cast Iron. S. N. A. Narayan and S. Visvanathan. (*Trans. Indian Inst. Met.*, 1955–56, **9**, 135–141). Attempts to produce nodular cast iron by the reduction of high-magnesia slags in contact with molten iron or the use of calcium and barium salts are described. The trials failed to show any commercial possibilities.—D. H.

Factors Affecting the Mechanical Properties of Pearlitic Nodular Irons. G. N. J. Gilbert. (*Brit. C.I. Res. Assoc. J. Res. Dev.*, 1957, **6**, Feb., 484–497). All pearlitic nodular irons have high ductile-to-brittle transition temperatures which in the notched impact test are well above room temperature. It is considered that poor tensile properties will be obtained in such materials because they are brittle at room temperatures. At 100° C above the transition range all materials have normal good tensile properties. Normalizing a pearlitic nodular iron generally results in an increase in strength because the pearlite is refined.—B. G. B.

New Nodular Iron Process. (*Indust. Heating*, 1957, **24**, Jan., 60–62). Nodular iron may be produced commercially by inoculation of melts with a mixture of $MgCl_2$, NaCl and Ca_2Si. The residual magnesium content is $< 0{\cdot}03\%$.

Condition for the Production and Quality Properties of Nodular Grey Cast Iron. F. Varga, B. Körös, E. Chapo, K. Janossy, and R. Sima. (*Kohászati Lapok*, 1954, **9**, Aug., 186–192; Sept., 193–211). The development of nodular grey iron production in Hungary and other countries is reviewed. Numerous experiments in the Hungarian Iron and Steel Research Institute confirm the principle that a successful inoculation depends both on the low carbon content and on the superheating of the melt. An increasing steel scrap content in the charge decreases the carbon content in the melt. Simultaneously an increase in tensile strength was observed in case of properly performed inoculations. The inoculation effect of CaSi has always been found better than that of FeSi.—P. K.

Development and Uses of Spheroidal Graphite (Nodular) Cast Iron for Marine Applications, with Special Reference to Diesel Engines. A. G. Arnold and B. Todd. (*Brit. C.I. Res. Assoc. J. Res. Dev.*, 1957, **6**, Apr., 588–599). Piston materials are compared and mechanism and prevention of failure and the effects of temperature are discussed. Other parts are considered.

Finishing Spheroidal Graphite Iron Castings. (*Indust. Finishing*, 1957, **9**, Apr., 516). The properties of the material are reviewed and surface coating processes briefly considered.

Chalk Talk on Diaphragm Moulding. T. E. Barlow. (*Mod. Castings*, 1956, **30**, Sept., 24–27). The process is described in discussion form.

A Research on the Workability of Cast Iron with Spheroidal Graphite. Chang Tso-mei, Shu U-yung, G. Tze-ming and Kuoh Sheng-chuen. (*Acta Metallurgica Sinica*, 1956, **1**, (2), 165–188). Material as-cast and annealed was subjected to

tensile, compression, impact, torsion, upsetting and rolling tests. The best temperature was found (depending on stress and rate of working) between 700–1100° C, and the maximum permissible working for a single operation. Industrial scale tests were compared.

Some Features of the Metallurgical Properties of Steel Castings. W. J. Jackson. (*Brit. Found.*, 1957, **50**, Apr., 211–219). The effect of ductility on minor variations in manufacture, S.P., non-metallic inclusions, interdendritic segregation and microporosity are considered. Effects of chemical composition and heat treatment on impact properties are then reviewed. (25 references).

We Cast Steel in Ceramic Molds. E. G. Vogel and D. C. Ekey. (*Mod. Castings*, 1956, **29**, May, 116–118). A report on Lebanon Steel Foundry's Ceramicast process is presented. Sillimanite ceramic is used and the process is compared with green sand, investment and shell processes.

Steel Castings Having High Creep Strength in the Medium and High Temperature Range. H. Zeuner. (*Giesserei*, 1957, **44**, Jan. 3, 1–7). The application of the characteristic value of the material and of the safety factors to the calculation of wall thicknesses is discussed. Creep-resisting materials for the various temperature ranges are reviewed. The time-dependence of the creep strength of typical cast materials in the medium and high temperature ranges is described. Finally the heat treatments necessary to produce these properties are given.—R. J. W.

Production of Corrosion Resisting Steel Castings for the Uranium Industry. W. G. Boustred. (*South African Mech. Eng.*, 1957, **6**, 199–211). Production of pump, filter and valve components is described with heat-treatment and notes on machining. High-frequency heating is used and sand or shell moulds, the usual material corresponding to A.I.S.I. 316.

Stickiness of Moulding Sand. K. Murata. (*Reports Government Ind. Res. Inst., Nagoya*, 1956, **5**, Nov., 544–547; summary 63). " Stickiness " is the tendency to adhere to pattern surfaces. It increased with moisture content and decreased with increase of clay and slightly with ramming force. Additions of silica flour, corn starch and bentonite reduce it.

Moisture Measuring on a Small Budget. G. DiSylvestro. (*Mod. Castings*, 1956, **29**, June, 57). A home-made device for sand control is described.

New Test for Better Molding Practice. R. W. Heine, E. H. King, and J. S. Schumacher. (*Mod. Castings*, 1956, **29**, Apr., 78–80). A test for the response of sand to jolt moulding is described. Effect of moisture and other factors on hardness are determined.

Pneumatic Reclaimer Lengthens Life of Sand. A. D. Barczak. (*Mod. Castings*, 1956, **29**, Feb., 24–27). The installation is described and properties of new and reclaimed sands are given.

Surface Flowability—Key to Smoother Castings. H. H. Fairfield and J. McConachie. (*Mod. Castings*, 1956, **30**, Sept., 30–32). A sand flowability test is correlated with casting smoothness and causes of poor flowability are discussed. Control to constant flowability has been attained.

Looking Into the Heart of a Mold. V. D. Frechette. (*Mod. Castings*, 1955, **28**, Nov., 58–62). The use of the polarizing and stereoscopic microscopes in the examination of foundry sand texture is described. Minerals are tabulated and their optical characteristics given, and clay minerals are included. (25 references).

Does Sand Testing Give Us the Facts ? R. W. Heine, E. H. King, and J. S. Schumacher. (*Mod. Castings*, 1956, **29**, Jan., 30–33). The effects of ramming and jolting are simulated on the laboratory scale. A modified ram procedure is used.

How Do You Store Your Patterns ? E. J. McAfee. (*Mod. Castings*, 1956, **30**, July, 52–56). The Puget Sound Naval Shipyard method of storage and record filing is described.

Glass Cloth Gives Old Patterns New Life. T. L. Arzt. (*Mod. Castings*, 1956, **29**, Apr., 72–73). A method of repair is given.

Contribution to the Problem of Increasing the Yield of Castings. W. Trommer. (*Giesserei*, 1957, **44**, Jan. 17, 37–45). The importance of increased yield is first discussed. The yield is dependent on the number and size of the gates used. By reducing these, using directional solidification, the yield can be increased. The effect on the casting is discussed. The size of gates can be reduced by using exothermic materials for the feeder heads. Experiences of the author with different materials are described. Finally an example is given for computing exothermic feeder heads.—R. J. W.

Progress in Taming the Sprue. G. Ohira. (*Mod. Castings*, 1956, **29**, Mar., 54–57). Gating methods are discussed and illustrated and filling times, mean flow velocity, and static pressure heads are tabulated. Step gating is reviewed and trials quoted.

Choke That Gate. By Using Set Ratios You Can Balance Flow Throughout the System. C. V. Knobeloch. (*Mod. Castings*, 1956, **30**, Dec., 48–50).

One Answer to the Snotter Problem. C. A. Sanders. (*Mod. Castings*, 1956, **29**, May, 66–68). Gating systems and other means of removing slag inclusions are discussed.

Investigation on the Behavior of Williams-Type Feeders as Down-Gates for Steel Castings. H.-U. Doliwa. (*Giesserei*, 1957, **44**, Feb. 28, 129–133). The principles of feeding techniques, including Williams-type feeders are discussed. The advantages and disadvantages of the Williams feeders are then enumerated. The special conditions when a combination of Williams feeders and open gates at atmospheric pressure can be used to advantage are described.—R. J. W.

Damage to Machines in Foundries. O. Neuhoff. (*Giesserei*, 1957, **44**, Jan. 31, 82–85). Causes of damage to machines used in the foundry are first discussed in general terms. Examples of damage are described and illustrated. These have been caused by foreign substances, material defects or by faulty manufacture. Examples are given of damage caused by neglecting regular maintenance e.g. scuffing.—R. J. W.

The Hand-Scraper in the Foundry. W. Gesell and W. Liege. (*Giesserei*, 1957, **44**, Feb. 14, 107–110). Various types of hand scrapers are first described and the possible uses of this type of device are enumerated. Special combinations of hand scrapers with other loading or handling mechanisms are illustrated. These are of great use in the small mechanized foundry for moving, in the first instance, sand and coal.

Mould Drying. Sub-committee TS39 of Technical Council, Inst. Brit. Foundrymen. (*Brit. Found.*, 1957, **50**, Apr., 169–183). It is found that moisture can move both towards and away from the heated surface and the portion driven back forms bands of high moisture content: that temperature (within the range 85–400° C) is the greatest single factor except in pressurized systems; that dried depth depends on moisture/temp./time relationships; pore space has more influence than permeability and the above apply to radiant, hot air circulating and portable drying methods. Pressure drying is also considered.

Dextrose Binder Speeds Core Production. C. F. Lourich. (*Iron Age*, 1957, **179**, Apr. 4, 104–105). Replacing conventional core oils with a dextrose binder has speeded up core production in several American foundries. The cores require 40–50% less time to bake through, and bake harder resulting in less metal penetration. Other advantages are briefly discussed and practical details are given.—G. F.

Which Core Process. W. B. Bishop. (*Foundry*, 1957, **85**, May, 180–183). Conventional, shell, air-setting and gas-setting core processes are discussed.

New Pointers on CO_2 Molding. (*Iron Age*, 1957, **179**, Apr. 25, 99–101). Tests on binders and charts for compressive strength as a function of treating time, type of binder and binder content are given.

Antifreeze Techniques for CO_2 Process. R. McIlrath. (*Mod. Castings*, 1957, **31**, Jan., 36–37). Prevention of solid CO_2 formation is discussed.

Pattern and Permanent Mould Construction at the GIFA Exhibition. F. Lamm. (*Giesserei*, 1957, **44**, Jan. 3, 19–25). The exhibitors at GIFA are listed and the DIN-specifications for wood, metal, and other materials given, together with a table of tolerances. Different qualities of patterns are demonstrated in this article which is amply illustrated. The mechanization of the construction of difficult patterns is described. Finally, various machines are described for the construction of permanent mould assemblies.—R. J. W.

Precision Castings in a Bigger Way. A. Dunlop. (*Mod. Castings*, 1956, **30**, Nov., 50–51). The ceramic mould process is briefly described, with examples.

Some Observations on the Production of Large Iron Castings. Z. G. Litostroj. (*Fonderia Ital.*, 1957, **6**, April, 153–156). [In Italian]. The author describes the tests carried out on samples taken from a large 12/14-ton propeller to assess the quality of the material in different parts of the piece. Both mechanical and micrographic analyses were carried out.

Large Iron Castings in One Branch of the Chemical Industry. G. A. J. Begg. (*Brit. C.I. Res. Assoc. J. Res. Dev.*, 1957, **6**,

Apr., 608–628). The alkali industry is considered and the effects of various defects in castings are reviewed. Recommendations are advanced.

Castings in Electrical Switchgear. C. A. C. Hillyer. (*Brit. C.I. Res. Assoc., J. Res. Dev.*, 1957, **6**, Apr., 600–607). The electrical properties of cast iron are also discussed.

Investment Castings: 4% Carbon Shot Controls Melt Chemistry. W. F. Carn and D. F. Mika. (*Mod. Castings*, 1955, **28**, Nov., 34–35). Control of carbon content by adding shot containing 4% C and other metals suitable for low-alloy, constructional or 400 series stainless or for Cr–Ni alloys including 300 series stainless is described.

Try Shell Molds For Low Volume Casting. S. Freedman. (*Iron Age*, 1957, **179**, Feb. 28, 84–85). The author discusses cases in which shell moulding, normally a mass production technique, can be economically applied to production of relatively small numbers of parts.—G. F.

Shell-Mould Castings for Complex Shapes, Smooth Surfaces, and Intricate Coring. (*Machine Design*, 1957, **29**, Mar. 7, 88–93). A profusely illustrated account taken from a forthcoming book.

Batch Melting Cast Iron Rotary Furnaces for the Shell Moulding Process. (*Mech. World*, 1957, **137**, Apr., 163). A 1-ton capacity unit is described and its consumption of power, fuel and refractories is given.

Soluble Wax Cores Solve the Impossible. C. W. Schwartz. (*Mod. Castings*, 1957, **31**, 56–57). The use of polyethylene glycol, 2 parts by weight and 160-mesh powdered mica, 1 part for core making is described.

New Shell Machine Cuts Resin Costs. R. S. Amala, J. H. Smith, A. L. Boegehold, and R. F. Thomson. (*Mod. Castings*, 1956, **29**, May, 62–65). General Motors Corp. blow-hot press moulding machine is described.

Shell Mould Castings Need Cleaning Too. E. F. Anderson. (*Mod. Castings*, 1956, **29**, Apr., 83). Variations in blasting are recommended.

Shell Molding Makes Better Cast Crankshafts with Lower Foundry and Machining Costs. H. N. Bogart. (*S.A.E.J.*, 1957, **65**, May, 26–27).

Cooperation Between Designer and Foundryman. M. M. Hallett. (*Brit. C.I. Res. Assoc. J. Res. Dev.*, 1957, **6**, Apr., 530–545). Design as related to characteristic defects in grey cast iron and to factors in production is discussed, and specific examples are then quoted of improvements effected in small castings.

Control of Shrinkage Defects in Cast Iron. J. Allan. (*Canad. Metals*, 1956, **19**, Nov., 36–42; Dec., 36–40). A detailed review of factors causing shrinkage defects in iron castings and methods of reducing shrinkage are presented.—B. G. B.

Contribution to the Criticism of Methods of Calculating Shrinkage Heads for Steel Castings. N. K. Ipatov and V. A. Fateev. (*Liteĭnoe Proizvodstvo*, 1956, (3), 27–32). [In Russian]. Methods for calculating shrinkage heads recently proposed by three authors are critically discussed and the need for a sufficient experimental basis for such calculations is indicated.—S. K.

Dynamics of the Solidification of Castings. I. D. Semikin and E. M. Gol'dfarb. (*Liteĭnoe Proizvodstvo*, 1956, (2), 16–22). [In Russian]. In this theoretical discussion of casting solidification a non-classical treatment of heat flow is adopted. In this, a given layer starts to take part in the flow process only after a time interval which increases with the distance of the layer from the surface. The discussion includes the dynamics of the growth of the crust with a constant temperature of the casting surface; the influence of the mould material on the temperature gradient in the crust and of a gap or facing layer on crust formation; and the calculation of mould thickness, worked examples being given.—S. K.

Analytical Solution of the Simplest Problems on the Solidification of Castings. N. G. Girshovich and Yu. A. Nekhendzi. (*Liteĭnoe Proizvodstvo*, 1956, (3), 14–19). [In Russian]. After a general discussion of the solidification of castings and the classification of this process into four types, the authors go on to develop equations for the solidification of flat castings. ((6), 15–18). In this instalment of their mathematical analysis of casting solidification the authors consider solutions of their equation for a sphere and for a cylinder without taking into account the physical heat of the solid metal.—S. K.

The CO_2 Process of Producing Moulds and Cores. With Reference to Stainless and Heat Resisting Steels. C. Sargent. (*Brit. Steelmaker*, 1957, **23**, July, 202–205). A review, with a general account of procedure and materials.

Technology for the Production of Moulds for the Centrifugal Casting of Pipes. V. M. Pishchev. (*Liteĭnoe Proizvodstvo*, 1956, (4), 10–12). [In Russian]. Defects in cast iron moulds for centrifugal casting of pipes are discussed and the influence of various factors on mould and product quality is discussed. Among factors considered are the composition of the cast iron, temperature and tube diameter.—S. K.

Grey Iron Castings. R. J. Fabian. (*Mat. Methods*, 1956, **44**, Oct., 121–136). Design data for grey cast iron are presented in a comprehensive review. Mechanical, physical and chemical properties, surface finishes, heat treatment, and welding and brazing are considered.—A. D. H.

Analytical Solutions of the Simplest Problems on the Solidification of Castings of Different Configurations. N. G. Girshovich and Yu. A. Nekhendzi. (*Liteĭnoe Proizvodstvo*, 1956, (4), 13–17). [In Russian]. This is a continuation of the theoretical study reported in *ibid.*, (3), 1956. The solidification of cylindrical and spherical castings is examined with the aid of previously deduced and newly derived equations.—S. K.

Chemical and Mechanical Requirements of Standard Investment Castings. (*Mat. Methods*, 1957, **45**, Jan., 143–145). The composition and mechanical properties of carbon, alloy steel and non-ferrous alloys standardized by the Investment Casting Institute are presented.—A. D. H.

Production of Accurate Heavy Castings. N. G. Romanovskiĭ. (*Liteĭnoe Proizvodstvo*, 1956, (3), 5–6). [In Russian]. The use of water-glass base mould and core mixtures, hardened by reaction with CO_2 is described. The relatively small extent, especially compared to Great Britain, to which such techniques are used in the U.S.S.R. is indicated.—S. K.

The Commonly Investment-Cast Alloys. (*Precision Met. Mold.*, 1957, **15**, June, 75–76, 78). Properties of steels and Cu and Al alloys are tabulated.

How Castings Hold Dimensional Stability. F. C. Benz. (*Iron Age*, 1957, **179**, May 16, 120–121). Tests showing the dimensional stability of Meehanite castings are reported, 9 in. bars changed less then 0·0001 in. in 7 years.—D. L. C. P.

Shell Castings: Typical Examples of Their Applications. H. G. Sieggreen. (*Mech. Eng.*, 1957, **79**, June, 560–562). A brief account of the process is given with illustrations and annotations of typical products.

Shell Mold Casting Solves Core Problem. (*Precision Met. Mold.*, 1957, **15**, June, 43, 46). Hydraulic circuit castings of intricate design were shell cast in grey iron.

Making Shell Moulds by a Chemical Hardening Method. B. Yu. Feĭgel'son. (*Liteĭnoe Proizvodstvo*, 1956, (2), 3–5). [In Russian]. The binder in the shell-mould production method described is silica gel. This is liberated from the sand–water–glass mixture by the action of CO_2. Excellent moulds are said to result.—S. K.

Experience in the Use of Ceramic Moulds and Shell Cores for Casting Large and Thin-Walled Parts. S. V. Russiyan and V. N. Bogdanov. (*Liteĭnoe Proizvodstvo*, 1956, (6), 4–7). [In Russian]. A procedure has been developed and tested in which large and thin-walled objects, such as cast-iron baths, are cast with the use of long-service ceramic moulds and bakelite-base shell cores. The introduction of the new technique in place of the ordinary mould and core practice reduced mould-mix consumption to 1/12 or 1/15, the labour involved in mould preparation and casting removal to about 1/3 and for cleaning to 1/8–1/10. Time spent on maintenance and on operation of ancillary processes was reduced by 30–40% and there was great improvement in the working conditions at the foundry.—S. K.

VACUUM METALLURGY

Equipment for Sintering in High Vacua. M. Mamula and J. Vacek. (*Hutnické Listy*, 1956, **11**, (11), 654–661). [In Czech]. The design and characteristics of a high vacuum furnace for sintering Ta and other high-melting metals are described.

Vacuum Process: Precise Heat Treat, Brazing Method for Assemblies. J. J. Lawless, jun. (*Western Metals*, 1957, **15**, May, 47–49). Advantages of vacuum over H_2 or unit gas are given and a unit is described and illustrated.

Experience in the Vacuum Annealing of Hot-Rolled Transformer Steel. N. F. Dubrov. (*Stal'*, 1956, (9), 811–815). [In Russian]. Vacuum annealing practice for hot-rolled transformer steel at a Soviet works is described and the results are compared with those obtained when tunnel kilns were

used. A vacuum of 50–70 mm. Hg and a maximal temperature of 1180° C could be maintained with the installations used. Both magnetic and plastic properties were better with the vacuum method and the surface and quality of sheets were improved. Sheets 0·50 and 0·35 mm thick of steel produced in arc and O.H. furnaces were practically identical. Further improvements desirable in vacuum annealing practice are discussed.—S. K.

On the Quality Change of Molten Iron during Holding under Low Presure. I. M. Okamoto and R. Yoda. (*Nippon Kinzoku Gakkai-si*, 1955, **19,** July, 397–401). [In Japanese]. The solidification behaviour of molten iron under vacuum at 1400° C shows three stages, corresponding with removal of dissolved O_2, removal of non-metallic inclusions, and appearance of massive or spheroidal graphite. Structural details are given for each. In deoxidized iron, the removal of O_2 causes a change in the form of the graphite; Mg treatment removes the last remaining O_2 and S and also forms a C–Mg bond which produces spheroidization of the graphite. (24 references).—K. E. J.

Quartz Vacuum Furnace with Molybdenum and Tungsten Heaters for Temperatures up to 3000° K. E. I. Smagina, V. S. Kutsev, and B. F. Ormont. (*Zavosdkaya Laboratoriya*, 1956, **22,** (10), 1249–1250). [In Russian]. A furnace in which temperatures of 1000° K and 2600° K have been obtained using respectively molybdenum and tungsten heaters is described. The respective constant-temperature volumes are 1·3 and 0·5 cm³.—S. K.

REHEATING FURNACES AND SOAKING PITS

Improvements of Operation by Remodelling of the Triple-fired Ingot-reheating Furnace. T. Otsuka, H. Matsumoto, and D. Yoshinari. (*Tetsu to Hagane*, 1956, **42,** Dec., 1106–1111). [In Japanese]. A triple-fired furnace for heating ingots of cross-section 400 mm or above was rebuilt with increased space at the bottom-heating zone, altered skid pipe construction, automatic control etc. Improvements were the reduced slag accumulation, reduced temp. gradient in the ingot and increased thermal efficiency. For future improvements, the number of skids should be small and the width large, to ensure adequate bottom heating; the bottom heating zone should be large, and large-capacity burners directed towards the middle of the ingots would reduce end burning.

Compartment Kilns for Heating Large Billets. B. S. Denezhkin. (*Stal'*, 1956, (10), 935–938). [In Russian]. The reconstruction of a Soviet-design furnace for heating billets from 700–1100 to 1220–1280° C is described. The top and bottom of the furnace as well as the gas burner arrangements were modified.—S. K.

Reconstruction of Gas Burners in Reheating Furnaces. M. N. Mel'nikov, Ya. S. Grodskii, and R. A. Bakshi. (*Stal'*, 1956, (11), 1035–1036). [In Russian]. A reconstruction of burners involving the adoption of axial gas feed, larger mixing chambers, and other measures is said to have greatly improved fuel economy at the Druzhkovskii works.—S. K.

Cascade Temperature Control System on a Billet Heating Furnace. J. M. Hess. (*Iron Steel Eng.*, 1957, **34,** May, 132–137). This paper describes an electric cascade control system which automatically adjusts the furnace control indices of a two-zone continuous billet heating furnace to make proper allowance for the various rates of mill operation. The system is installed on the 14 in. mill furnace at Bethlehem Steel Co.—M. D. J. B.

Experience in the Application of Automation of Rolling-Mill Heating Installations. A. M. Kulakov and I. M. Gel'fand. (*Stal'*, 1957, (1), 80–83). [In Russian]. The automatic regulation of regenerative soaking pits and continuous furnaces and the effects its adoption has had on rolling practice at Magnitogorsk are described. The system for the pits embraces regulation of pit temperature, of gas/air ratio and of pressure in the pit working space as well as valve reversals and the stopping of gas and air flows. For temperature measurement in both pits and furnaces optical pyrometers were found to be preferable to thermocouples. The benefits of automation are clearly shown by comparison of metal temperature measurements after the roughing group, and extension of the scope of automation is urged.—S. K.

Use of Protective Coatings for Reducing Metal Decarburization. S. Z. Yudovich. (*Stal'*, 1957, (1), 69–71). [In Russian]. Experiments on the use of brushed-on water-glass coatings to reduce the deoxidation of billets of carbon tool, ball-bearing and spring steels during heating are described. Not only did this technique make scale more easily removable, but together with improved heating arrangements, reduced decarburization rejects to 1/10 of their former value.—S. K.

Continuous Rapid Heating of Tubes Before Reduction. N. Yu. Taĭts, G. N. Kheĭfets, and V. B. Tov. (*Stal'*, 1956, (9), 826–830). [In Russian]. The disadvantages of present types of furnaces for the pre-reduction heating of tubes, conditions for rapid heating and the design and positioning of continuous furnaces are discussed. Calculations indicate that potentially great advantages could be obtained from continuous furnaces.—S. K.

Elimination of Axial Burning of Rail-Steel Ingots. M. P. Sidel'kovskiĭ, L. P. Tarasova, and O. V. Dolinenko. (*Stal'*, 1956, (10), 915–920). [In Russian]. Macrostructural investigation of rails at the "Azovstal'" works sometimes revealed zones of increased etchability. Such rails were rejected and a comprehensive investigation, described in this article, was undertaken to find the causes. From a study of both rails and blooms it was established that this type of defect is due to axial burning due to excessive ingot heating temperatures, leading to local saturation of the fused zones with O_2, N_2 and H_2. It was also shown that satisfactory external appearance of the blooms is no guarantee of absence of burning. By careful control of heating temperatures the occurrence of the defects was avoided.—S. K.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Recent Heat Treatment Furnace Installations. Annual Survey of Developments. (*Metallurgia*, 1957, **55,** June, 275–294). Reheating, stress-relieving, case-hardening, hardening and tempering, salt-bath, general utility and miscellaneous furnaces and induction heating installations are described in an illustrated review.

Applications of the Shaker Hearth Furnace. W. H. Statham. (*Metalworking Prod.*, 1956, **100,** Oct. 28, 1730–1733). Furnaces for the heat-treatment of small carbon steel components are described. First used for needles they have been adapted to springs and similar parts. Continuous conveyor quench tanks are fitted, and protective atmospheres are used.

G.E.'s New Industrial Heating Plant at Shelbyville, Indiana. (*Indust. Heating*, 1957, **24,** Feb., 275–290, 426–430). A description is given of the lay-out and products of a new, large furnace-building establishment in U.S.A.

Basic Furnace Atmospheres and Composition Effects. C. H. Vaughan. (*Indust. Heating*, 1957, **24,** Feb., 248–258, 424). The generation and properties of exothermic, and endothermic gas, N_2, dissociated ammonia, H_2, A and He are discussed.—A. D. H.

Safe Operation of Atmosphere Furnaces. J. Huebler. (*Steel Processing*, 1957, **43,** June, 337–340, 342–343). Carbon monoxide hazards, explosion conditions (combustion, ignition and explosive limits) and purging operations are briefly discussed and emergency precautions with a chart of protective systems, safety devices and equipment and the responsibility of the user are then considered.

Installations for Providing a Protective Atmosphere by the Hydrogen Combustion Method. G. K. Chernov. (*Stal'*, 1956, (8), 755). [In Russian]. At a works without a gas supply protective atmospheres were produced by using electrolytic hydrogen to remove the oxygen from air. The nitrogen was used in annealing cold-rolled transformer steel after mechanical working.—S. K.

A Three-Line Heat Treatment Station. (*Metalworking Prod.*, 1957, **101,** May 3, 757–758). An account of David Brown Industries gearbox division Huddersfield plant.

Heat Treatment of Transmission Gears. (*Metallurgia*, 1957, **55,** May, 232–234). A new plant for processing 1000 lb of commercial vehicle transmission gears every 7 hrs. is described. It is installed at the Park Works, Huddersfield, of the Automobile Gearbox Division of David Brown Industries Ltd.—B. G. B.

Changes in Mechanical Properties of Steel Caused by Stress-Ageing Treatment. I. T. Sakurai, T. Kawasaki, and H. Izumi. (*Nippon Kinzoku Gakkai-Si*, 1955, **19,** July, 401–404). [In Japanese]. An apparatus is described in which a speci-

men loaded in an Amsler machine can be transferred to a furnace for stress-ageing. This treatment after cold-working increases the yield point. Treatment at 200° and 300° C produces zig-zag stress-strain curves. Holding time has little effect. When an annealed specimen is treated at a stress just below the yield stress, the upper yield point and magnetic permeability are markedly increased. Fatigue strength is also improved by the treatment.—K. E. J.

Influence of Heat Treatment on the Properties of Low-Carbon Rimming Steels. M. V. Pridantsev and Kh. Sh. Levinzon. (*Stal'*, 1956, (11), 1006–1015). [In Russian]. In this article the factors making low-carbon steel liable to ageing are discussed and a heat treatment to counteract this tendency is described. This immunization, which applies both to mechanical and thermal ageing, could be achieved by rapid cooling from above A_3, and was then accompanied by marked improvement in properties (cold brittleness, yield point and strength). Some advantages to be expected from using the immunized rimming instead of killed steels are listed. The influence of vanadium, manganese and boron on the properties of low-carbon steels are also discussed, together with some theoretical aspects of the ageing process.—S. K.

High Quality Gears and Assembly Parts Require Precise Heat Treating. J. T. Mitchell and R. Buyea . (*Indust. Heating*, 1957, **24,** Feb., 260–268, 274). Annealing, gas carburizing and salt bath furnaces for the annealing and case-hardening of gears at the New Process Gear Corp., Syracuse, N.Y. are described.—A. D. H.

On Application of [Town's] Gas for Bright Heating and Carburising of Steels. III. On Bright Heating of Steels in Inert Gas Produced by the Combustion of [Town's] Gas with Air. S. Uchida. (*Nippon Kinzoku Gakkai-Si*, 1955, **19,** Sept., 517–520). [In Japanese]. Atmospheres containing high proportions of N_2 were produced by partial or complete combustion of town's gas in air, followed by absorption of CO_2 by a water spray and reduction of CO_2 and H_2O in a converting furnace. They were shown to have little attack on steel even at high temperatures. (12 references).—K. E. J.

Some Practical Aspects of Pack Carburising. (*Wild-Barfield Heat-Treatment J.*, 1957, **6,** June, 11–15). Steel compositions, an outline of the process, a table of time-case depths, and notes on furnaces and their operation are given.

Surface Hardening—A Review. W. L. Besselman. (*Metalworking Prod.*, 1957, **100,** Dec. 21, 2088–2092). Carburizing, nitriding, carbonitriding, cyaniding and induction hardening are included, structures are shown and mechanized furnaces described, including salt-bath types.

Flame Hardening. (*Torch*, 1957, Spring, 2–7). An illustrated review.

Shop-Made Flame-Hardening Machine Improves Shaft Quality. P. J. Bruno. (*Metalworking Prod.*, 1956, **100,** Oct. 28, 1134–1735). Shafts up to 8 in. dia. and up to 20 ft long can be treated with oxy-acetylene burners hardening to 57–62 Rockwell C.

Automatic Flame Hardening. (*Automobile Eng.*, 1957, **47,** June, 244–249). Machines for operation in the production line are described.

A Versatile Induction Hardening Equipment. (*Wild-Barfield Heat-Treating J.*, 1957, **6,** June, 6–8). An installation for small gears, shafts, etc., at Wolf Electric Tools Ltd., Ealing, is described.

Annealing Malleable Iron in a Liquid. V. N. Smorodinskiĭ. (*Liteĭnoe Proizvodstvo*, 1956, (6), 31). [In Russian]. A two-stage annealing of malleable-iron is briefly described. The first stage is carried out in a bath of fused NaCl (70% by volume) and $BaCl_2$ (30%), giving a structure of globular carbon + lamellar pearlite.—S. K.

Advantages and Disadvantages of Flame Annealing from the Point of View of its Effects on the Strength of the Metal in the Welded Zone. H. Pohl. (*Zváranie*, 1956, **5,** (7), 194–197). [In Czech]. Studies relating to the strength of welds carried out in the German Democratic Republic, particularly in relation to problems arising in the construction of welded ships, are briefly described.—P. F.

Dimensional Changes due to Annealing of Plastically Cold-worked Metals. V. On the Effect of the Microstructure of Eutectoid Steel Bars. VI. Of Eutectoid Steel Bars Compressed after Elongation. H. Sekiguchi and M. Inagaki. (*Nippon Kinzoku Gakkai-Si*, 1955, **19,** Aug., 494–497; 498–501). [In Japanese]. V. Dimensional changes due to annealing of eutectoid steel bars containing lamellar pearlite are in the opposite direction to deformation due to cold-working before annealing. With those containing granular pearlite, they are generally in the same direction. The tendencies to hardening due to cold-working and the elastic after-effect are greater in the case of lamellar pearlite. VI. Dimensional changes due to annealing of eutectoid steel bars compressed to the original dimensions after elongation resemble, but are smaller than, those due to annealing of compressed specimens.—K. E. J.

On the Change of the Modulus of Rigidity in 18–8 Stainless Steel Wires due to Cold Drawing and Annealing. H. Takei. (*Nippon Kinzoku Gakkai-Si*, 1955, **19,** July, 448–452). [In Japanese]. At the early stage of cold drawing, the modulus of rigidity decreases markedly, probably because of formation of α-phase. For annealing in the ranges 100–250°, 400–600° and 700–800° C, the modulus increases, as the result of strain hardening, crystal recovery and recrystallization, respectively. Intergranular corrosion is found in specimens annealed at 600–700° C.—K. E. J.

On the Deformation of Self-hardening Tool Steel due to Heat-treatment. IV. Effect of Cyclic Treatment. K. Yokoyama and T. Matsukura. (*Nippon Kinzoku Gakkai-Si*, 1955, **19,** Sept., 513–517). [In Japanese]. Specimens were prepared of 2% C, 13% Cr steel and 1% C, 5% Cr steel, quenched in air from 1130° C to retain much of the austenite. Six cycles of cooling to − 75° C and tempering at 100–600° C were performed, followed by ageing at room temp.; length measurements were made to examine the stability. Tempering at 300° C produced the maximum stabilization of residual austenite. Ageing produced small changes in length, except when the tempering temp. had been approximately 500° C.

The Transformation of Ball-bearing Steel. K. Abe and R. R. Hasiguti. (*Tetsu to Hagane*, 1956, **42,** Dec., 1123–1127). [In Japanese]. The continuous cooling and isothermal transformation of ball-bearing steel after austenitizing at 1050° C for 30 min were studied. The calculated curve for the beginning of pearlite transformation during continuous cooling agreed reasonably well with that found experimentally. No primary cementite and bainite were found in the products of continuous cooling transformation with higher contents of carbon and chromium, but they were with lower contents.

Stepped Austenitizing Treatment for 4340 Steel. E. P. Klier, V. Weiss, and G. Sachs. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209;** *J. Met.*, 1957, **9,** Apr., Section 2, 424–425). The authors have studied the effects of stepped-austenitizing during the heat treatment of 4340 steel. They conclude that deeper hardenability may be promoted, with reduced notch sensitivity in the high-strength range.—G. F.

On the Mechanical Properties of Semi-rapid Heated and Quenched Steel. II, III. S. Owaku and K. Iijima. (*Nippon Kinzoku Gakkai-Si*, 1955, **19,** July, 408–411). [In Japanese]. Tests on a 0·6% C steel show that the greatest toughness is acquired by quenching austenitized steel of which all primary ferrite has just been dissolved into austenite, whatever the method used for austenitizing; this explains the superior toughness of semi-rapid heated and quenched steel. Wear resistance is improved by quenching whether by this or conventional methods.—K. E. J.

On the Deterioration of Quenching Oils. III. Some Experiments on the Deterioration Inhibitor. M. Tagaya, I. Tamura, and K. Sugimoto. (*Tetsu to Hagane*, 1957, **43,** Jan., 61–66). [In Japanese]. Many oxidation inhibitors apparently prolong the initial induction period of deterioration of the oil, but undesirably accelerate the later polymerizing period by catalytic action. Good effects, however, were observed with *N*–phenyl–1–naphthylamine in soya-bean oil, diphenylamine and di–*tert.*–butyl–*p*–cresol in 110 dynamo oil and in 120 cylinder oil.—K. E. J.

The Tempering of Low-alloy Creep-resisting Steels containing Chromium, Molybdenum, and Vanadium. E. Smith and J. Nutting. (*J. Iron Steel Inst.*, 1957, **187,** Dec., 314–329). [This issue].

Studies on Tempering of Carbon Steel Samples by Thermo-electromotive Forces. J. K. Mukherjee and G. P. Chatterjee. (*Trans. Indian Inst. Met.*, 1954–55, **8,** 217–222; discussion 222–224). Measurements of thermo-e.m.f. of quenched 0·8% C steel tempered at 50° and 75° C showed that the e.m.f. falls to a minimum value followed by an increase.

Structural Changes Occurring in Low-Carbon Alloy Steels on Tempering. J. Wozniak and V. Foldyna. (*Hutnické Listy*, 1956, **11,** (11), 663–669). [In Czech]. Dispersion hardening in two steels containing each about 0·1% C, and

1·24% V and 2·25% Mo respectively, was studied by electron-diffraction at various stages of tempering at temperatures ranging from 350 to 650° C. Maximum hardness was found to be attained prior to the detectable presence of V_4C_3 and Mo_2C and Fe_3C, i.e. detectable precipitates exceeded the size characteristics for critical dispersions. Certain divergencies between the results now obtained and those of other authors working with similar steels are discussed.—P. F.

Study on Timken 16–25–6 Type Heat-resisting Alloy. III. Change of the Properties during Tempering and Precipitates. Y. Imai and K. Tanosaki. (*Tetsu to Hagane*, 1956, **42,** Dec., 1132–1138). [In Japanese]. Changes of the properties during tempering (0–1000° C) of the alloy solution-treated at 1250° C were followed, and the microconstituents of the alloy aged between 550° and 900° C were examined. During tempering a precipitate was observed at 500°–550° C, which dissolved, and there were at least two further precipitations in the range 700–1000° C, with a transition temperature at about 825° C. Below 800° C, the microconstituents were [(Cr, Fe, Ni)$_3$Mo]C, and above 800° C [(Cr, Fe, Ni)Mo]N and [(Cr, Fe, Ni)$_5$Mo]C. No evidence of sigma phase was found.

FORGING, STAMPING, DRAWING, AND PRESSING

Forging High-Speed Steel. R. Suchopárek. (*Hutník*, 1956, **6,** (10), 304–308). [In Czech]. The technology of forging high-speed steels is described, from the ingot stage to the final product.—P. F.

Automatic Forging Without Special Equipment. H. Chase. (*Iron Age*, 1956, **178,** Nov. 15, 138–139). Mechanization of loading and handling on a standard upsetter is described. Automatic transfer and induction heating was arranged.

Designing a Forged Steel Crankshaft. H. F. Wood. (*S.A.E.J.*, 1957, **65,** June, 23–25). An automobile crankshaft design is considered in terms of the properties of the material.

The Production Technology of Die-Forging. J. Šedivec. (*Hutník*, 1957, **7,** (3), 94–100). [In Czech]. The design of dies and correct forging methods are discussed.—P. F.

The Production of Hollow Boiler Shells from Hollow Ingots. O. Bohuš. (*Hutník*, 1956, **6,** (10), 299–301). [In Czech]. The forging of boiler shells from 100-ton ingots is described on the basis of the author's experience of Soviet developments in this field.—P. F.

Manufacture of Large Composite Wheels with Rims in LCN–155 Steel. Y. Ishihara, S. Tamaki, and K. Deguchi. (*Tetsu to Hagane*, 1957, **43,** Feb., 146–153). [In Japanese]. Details are given of experimental work and production methods used for making wheels, with rims forged from LCN–155 super-high grade heat-resisting steel, of outer diameter 600 mm, inner diameter 300 mm and thickness 60 mm, with Cr–Mo steel bosses. The best welding results were obtained with 5 mm rods of the same LCN–155 steel, using the argon-arc tungsten method. The rim of the wheels could withstand 750–800° C, and $\frac{1}{3}$–$\frac{1}{4}$ of the expensive heat-resisting steel was saved by the composite construction.—K. E. J.

Spray Coating for the Protection of Sheet During Forming. (*Machinery*, 1957, **91,** July 12, 76–77). A synthetic resin coating material is described.

Cold Headed Parts from Austenitic Stainless Steels. S. E. Tyson. (*Mat. Methods*, 1956, **44,** Oct., 102–106). The selection of austenitic Cr–Ni stainless for cold heading is considered in terms of corrosion resistance and strain hardening. Wire quality and part design are also discussed.—A. D. H.

Explosive Forming Socks Tough Alloys in Shape to Thwart Thermal Thicket. (*Western Metals*, 1957, **15,** May, 57–58). Explosive forming processes developed by Lockheed Aircraft are described. Both propellants and high explosives can be used and a few details are given of uses and of the forces involved.

Production of Non-Untwisting Cables. V. I. Sukhinin. (*Stal'*, 1956, (10), 924–927). [In Russian]. Tests on non-untwisting steel cables have shown that as good a performance is given by those made with preliminary deformation only of the strands as those in which the wires were previously deformed as well.—S. K.

Recent Developments in Wire Rope. T. Nishioka. (*Tetsu to Hagane*, 1957, **43,** Jan., 67–76). [In Japanese]. A review, covering steel compositions, calculations of mechanical properties, fracture data, testing methods and results, lays and special ropes, e.g., non-spinning and non-rotating ropes.

Mechanical Properties of Cold-Worked Steel Wire. K. I. Tulenkov, M. I. Zlotnikov and S. F. Bobyleva. (*Stal'*, 1956, (9), 821–825). [In Russian]. Empirical formulæ and nomograms are presented for the rapid calculation of the mechanical properties of carbon steel wire. Good agreement between calculated and actual results is reported.—S. K.

Studies on Helical Springs Made of Twisted Wire. I. Fundamental Phenomena. II. General Consideration on the Effect of Twisting. S. Mito. (*Nippon Kinzoku Gakkai-Si*, 1955, **19,** Aug., 453–456; 456–460). [In Japanese]. I. The mechanism of pretwisting to obtain strong helical springs is examined. When loading, the coil diameter of a spring made of cold-drawn wire expands, while it contracts after annealing. Annealing also changes the length of the spring. These phenomena are explained in terms of anisotropic properties of the wire and the fading of residual stresses. (13 references). II. A remarkable improvement in strength is obtained by ordinary twisting during coiling, and a slight improvement by twisting before coiling. Inverse twisting produces a notable decrease in fatigue strength as compared with elastic limit, because of tensile stresses acting on crack planes. Prestressing ("setting") produces a reasonable increase in elastic limit, but little in fatigue limit. (11 references).

An Electrical Analogue for Estimating Die Temperature During Wire Drawing. A. E. Ranger. (*Wire Prod.*, 1957, **6,** May, 5, 7, 9, 11, 13, 15). Temperature outside the die pellet was measured and interior surface temperature in contact with the wire computed by a resistance network analogue. It is concluded that cooling of the front surface of the die is important.

Cold Extrusion of Steel on Verson Presses. (*Machinery*, 1957, **91,** July 12, 90–94). A brief account of the process and products.

Steel and Titanium Extrusions Today: Increasing in Demand Tomorrow: Unusual Shapes. F. T. Roberts, jun. (*S.A.E.J.*, 1957, **65,** June, 75–77). Achievements to date and possible future progress with stepped and similar shapes are outlined.

ROLLING-MILL PRACTICE

Metal Deformation Stress Conditions and Strip Rolling Factors. M. A. Benyakovskiĭ, V. I. Kulikov, V. A. Shadrin, I. P. Kolpakov, Ya. S. Kutuev, G. G. Kustobaev, M. F. Kochnev, I. V. Esipov, and B. I. Petrov. (*Stal'*, 1957, (1), 59–63). [In Russian]. An account is given of full-scale investigations in which roll-pressure, strip tension, power consumption for each stand, and rolling speed and thickness after each stand were determined simultaneously. The mill was the continuous strip cold rolling mill at Magnitogorsk and both carbon and special steels were rolled. Wire resistance strain gauges were successfully used as transducers and they are recommended for routine control of continuous and reversing cold rolling mills. Data on roll pressures and specific power consumptions are suitable for improving mill operation and for mill design.—S. K.

Automatic Metallization of Steel Rolls under a Ceramic Flux. K. V. Bagryanskiĭ, N. F. Protasov, V. A. Tyagus and Yu. A. Shapiro. (*Stal'*, 1956, (11), 994–997). [In Russian]. An account is given of a method of renovating worn steel rolls by electric-arc metallization under a layer of ceramic flux. The method developed at the Azovstal' works is applicable to any size of roll and is better than other Soviet methods.

The Design and Construction of Slabbing Mill Rolls. J. Sklenář. (*Hutník*, 1957, **7,** (1), 16–22). [In Czech]. The processes are discussed in detail, complete examples being worked out. The methods are based primarily on Soviet practice.—P. F.

Improving Mould-Construction for Cast-Iron Rolls. P. D. Mikhalev. (*Liteĭnoe Proizvodstvo*, 1956, (2), 23–24). [In Russian]. A self centering device has been developed which ensures that the journals and wobbler of cast-iron rolls are co-axial with the roll. The construction and the use of this equipment in mould assembly are described.—S. K.

[Electron-microscopical Structure] of the Work Roll for Cold Rolling. I. T. Ando, K. Gokan, and K. Arase. (*Tetsu to Hagane*, 1957, **43,** Feb., 127–132). [In Japanese]. The electron microscope can be used to examine the amounts of spheroidized carbides, martensite matrix and retained austenite at the surface layer of the roll. To obtain a properly spheroidized structure after forging, the most effective

annealing treatments are those in which there is phase transformation. Slower cooling rates after holding at 780° C give better carbide distributions. Repeated heating and cooling between 700 and 780° C gives larger carbides and a marked development of pearlite lamellæ.—K. E. J.

Practical Investigation on Work Rolls for Cold Rolling. Y. Kuwabara, Y. Miwa, and K. Ariga. (*Tetsu to Hagane*, 1957, **43**, Jan., 24—28). [In Japanese]. Investigations were made of 200 rolls, of 16½ in. diameter, used in four-high cold reversing mills. Rolls which operated well and gave good products showed a linear decrease of hardness with depth below the surface; bad rolls showed extreme softness 5–6 mm below the original surface. Electron-microscope studies showed that surface structures could be classified as: (i) martensite matrix with 0·5–2·0μ cementite particles, (ii) martensite matrix with net structure and cementite particles, and (iii) martensite and austenite. No changes occurred during operation, and group (i) rolls gave the best results.—K. E. J.

Modernization of Rolling-Mill Gear Housings. S. M. Ruvinskiĭ, I. S. Starets, and E. E. Garmarsh. (*Stal'*, 1956, (9), 849–851. [In Russian]. On the basis of practical experience the design of rolling-mill bearings is discussed and the substitution of ball for friction bearings at heavily loaded points is recommended.—S. K.

Type DTs Plastic—A New Substitute for Tekstolit and Non-Ferrous Metals. A. I. Neveduyuschchiĭ. (*Stal'*, 1956, (10), 940–941). [In Russian]. The plastic recommended for use in roll bearings is of the bakelite type, filled with paper fibres. Satisfactory test results at several rolling mills are reported.—S. K.

Increased Productivity of No. 2 Blooming Mill. V. P. Kozhevnikov, A. M. Uzienko, G. G. Kustobaev, G. V. Savel'ev, and F. P. Skachko. (*Stal'*, 1957, (1), 47–52). [In Russian]. Since 1946 the productivity of the 1150 No. 2 blooming mill at Magnitogorsk has increased by 86·5%, with a fall in rolling-defect rejects from 0·30 to 0·13% and a halving of the idle time. In this article the measures leading to these improvements are described, mainly the provision of two additional stands in front of the 630 mill and two (one horizontal—one vertical roll) between the blooming mill and the first shears, automation of auxiliary mechanisms, use of larger and more suitably shaped ingots and enlargement and improvement of heating and cooling facilities.—S. K.

Results of the Reconstruction of No. 3 Blooming Mill at Magnitogorsk. L. V. Andreyuk. (*Stal'*, 1957, (1), 53–59). [In Russian]. Until its reconstruction the bottlenecks on No. 3 blooming mill were the reversing 1150 stand, the feeding of hot ingots to the blooming mill and, to some extent, the soaking pits had to be eliminated, giving increases of 8·5–47% in the hourly productivity. The daily productivity has only risen by about 5%, mainly through insufficiency of heating facilities and the extension of these together with the use of a higher C. V. fuel and the elimination of newly emerged bottlenecks is planned.—S. K.

Rationalizing the Supply of Metal Rolled Products. M. I. Portnyagina and K. P. Murashkin. (*Stal'*, 1956, (10), 921–924). [In Russian]. Examples of the extent to which unnecessary transport of rolled steel occurs in the Soviet union are given and far greater decentralization is recommended. This, it is claimed, would enable users' requirements to be supplied more exactly, and a more equitable distribution of supplies between customers to be achieved with less transport. The present centralized organization Glavmetallosbyt, with its forty branches throughout the Union, is considered unequal to its proper tasks. Typical transport costs per ton of various types of steel in the U.S.S.R. are given.—S. K.

Comeback for the Universal Mill. (*Steel*, 1957, **140**, May 20, 146–148). Characteristics of the universal mill are considered; 3 layout plans are described. Demands for high production and bigger slabs give this type of mill an advantage over other ingot-breakdown types.—D. L. C. P.

Rolling Square-Section Steel with Sharp Edges. M. V. Shuralev and S. G. Nekrasov. (*Stal'*, 1956, (11), 997–999). [In Russian]. Roll-pass designs for rolling squares to produce sharp edges and flat sides is discussed. Separate designs for 8–10 mm and 11–40 mm are recommended.—S. K.

Contribution on the Rolling of Rails. B. V. Merekin. (*Stal'*, 1956, (9), 803–805). [In Russian]. The folding of the head when rolling rails in skew passes could be avoided by correct selection of the displacement of the centre of gravity of the pass from the rolling centre-line.—S. K.

On the Improvement of Roll Pass Design for Rails. A. M. Karpunin. (*Stal'*, 1957, (6), 536–540). Development of expansion stresses must be avoided and the best shape of ingot is rectangular, the head and foot being formed from the narrow faces. With deep knifing of the lower roll in the first trapezoidal pass, uneven deformation occurs in the foot and hair cracks may form. Increase of thickness of the bloom improves the rail surface, but it must be considered in relation to the shape of the ingot and the thickness of the columnar crystal zone. 2–3 T-shaped passes are recommended and to reduce the spread in the passes to 6–8 mm five plastically deforming passes should be used.

Influence of Rolling Factors on the Quality of Bessemer Steel Rails. N. I. Beda, N. K. Kotov, G. P. Borisenko, and E. M. Mal'tseva. (*Stal'*, 1956, (10), 897–900). [In Russian]. A large-scale investigation on the quality of Bessemer steels in relation to (*a*) temperature of charging ingots into the soaking pits, (*b*) ingot temperature at the start of rolling in the blooming mill, (*c*) rolling factors in the mill and (*d*) roll wear. All these were found to affect quality appreciably. For 4·36-tonne ingots a charging temperature of 910° C gave high soaking-pit productivity and low rejects; optimal start-of-rolling temperatures were over 1180° C. The degree to which intensification of reduction on the blooming mill could be effected depended on metal quality. Periodic treatment of rolls with a special cutter was required to avoid hair crack formation in the rail.—S. K.

Rolling of a Special Wide-Flanged Channel. M. M. Shternov. (*Stal'*, 1957, (1), 90–91). [In Russian]. Details of roll-pass designs and rolling factors used to produce a deep steel channel for railway wagon construction are given.—S. K.

Use of Roller Guides for Rolling Heavy Sections. M. V. Shuralev and P. I. Karyakin. (*Stal'*, 1957, (6), 548–549). Productivity increased by 4–6% with improvement in quality.

Rolling of Railway Tyres. A. Vach. (*Hutník*, 1957, **7**, (2), 47–50). [In Czech]. Modifications in heat treatment and rolling methods introduced in railway tyre production in Czechoslovakia are discussed, and the general production technology is considered.—P. F.

Possibilities of Improving the Quality of Railway Tyres by Manufacture from Round Rolled Billets. M. Žídek and O. Pejčoch. (*Hutnické Listy*, 1956, **11**, (12), 709–713). [In Czech]. Technological and economic aspects of the method are considered. Comparison shows both quality improvements and cost reductions in comparison with the conventional method of producing tyres from pear-shaped ingots.—P. F.

Experience in the Adoption of Mechanized Coolers in Section Rolling Mills. I. A. Lipovetskiĭ, A. A. Nefedov, P. M. Gratsershtein, D. P. Lobkovskiĭ, and V. A. Landa. (*Stal'*, (9), 798–802). [In Russian]. Modifications are described to a Soviet-designed cooler, which have enabled the operation of the flying shears and some other parts of a section rolling mill to be improved.—S. K.

Design and Performance of Hydraulic Descaling Units in Rolling Mills. W. Berns. (*Stahl u. Eisen*, 1957, **77**, May 2, 567–576). Descaling in three types of rolling mill—two continuous and a slab mill—is described. Pressures used were respectively 80, 100, and 125 atm., and angles of incidence were 15° and 18°. The hydraulic unit is described in detail and efficiency values, quantities of water used, etc., quoted. The degree of descaling obtained was sufficient for ordinary requirements, but really satisfactory results could not be attained. Descaling depends on the mode of operation of the rolling mill and the unit as well as on the properties of the steel. The results obtained were not consistent.—T. G.

Cleaning Alloy Steel Slabs Prior to Rolling. J. Kostkuba. (*Hutník*, 1956, **6**, (12), 366–371). [In Czech]. Modern methods of cleaning by mechanical means are considered, and the role of automation in the processes is discussed. Considerable savings in production costs are obtained in comparison with pickling.—P. F.

Rational Method of Production of Ingots of Sheets. I. D. Kuzema. (*Stal'*, 1957, (6), 541–543). Formulæ for relating ingot size to dimensions of sheets required are given.

Effect of Roll Shape and Temperature on the Weldability of Sheets in Packet Rolling. D. I. Suyarov and P. F. Zasukha. (*Stal'*, 1956, (10), 901–904). [In Russian]. In this article theories on the reasons for sheet-to-sheet welding during

packet rolling are critically reviewed and some theoretical views are advanced. As a practical measure it is recommended that, to minimize welding, the thinner the sheet being rolled, the higher should be the roll temperature and the lower the metal heating temperature.—S. K.

Improvement of the Production of High-Chromium Steel Sheet. A. A. Babakov, T. A. Zhadan, V. A. Danilin, S. F. Bakuma, K. I. Antipov, M. N. Kul'kara, and S. Z. Kupryakhina. (*Stal'*, 1957, (6), 555–559). Special rolling and heat-treatment schedules have improved the production of X25T and X28 steels, giving fine-grained structure with improved ductility and elasticity.

Study of Factors Influencing Surface Staining of Cold Rolled Steel. C. R. Lillie, D. W. Levinson, F. V. Schossberger, K. Hattori, and H. Marver. (*Iron Steel Eng.*, 1957, **34**, May, 69–79). The first part of this paper discusses types of defects on cold rolled steel and their occurrence, the second studies the reaction mechanism of black stain formation on annealed steel coils. (13 references).—M. D. J. B.

Calculation of the Power Requirements of Sheet Roller Straightening Machines. A Geleji and G. Devenyi. (*Acta Techn.*, 1957, **17**, (3/4), 361–380). The formula for work of deformation is deduced and power demand calculated. Measurements are then made and compared with the formula giving reasonable agreement.

Preset Screwdown Control of Reversing Rougher at Granite City Steel Co. A. W. Schlechte and C. A. Schurr. (*Iron Steel Eng.*, 1957, **34**, May, 109–115). The authors describe the automatic operation of the screwdown and edger adjustment on the reversing rougher of Granite City Steel Co's 90 in. hot strip mill. Operation of the equipment is described as well as the schedule of the selector panels.—M. D. J. B.

Mill Setup Controls, Records Strip Thickness. (*Iron Age*, 1956, **178**, Nov. 15, 140–141). X-Ray gauging with automatic control of screwdown and back tension is arranged for rolling stainless strip 0·005–0·040-in. thick.

Measuring Hot Steel Strip Thickness. (*Metallurgia*, 1957, **55**, May, 241–243). A γ-ray thickness gauge developed by the Baldwin Instrument Co. Ltd. and installed at Steel, Peech, and Tozer Ltd. at Rotherham is described.—B. G. B.

Weirton Steel Company's New No. 7 Tandem Cold Mill. C. F. Buente. (*United Effort*, 1957, **37**, Second Quarter, 3–8).

Production of Cold-Rolled Sheet for Dynamos. N. I. Sheftel'. (*Stal'*, 1957, (6), 562–563).

Modern Techniques in the Cold Reduction of Sheet and Tin Plate. E. N. Archibald. (*Iron Steel Eng.*, 1957, **34**, May, 141–146). The author describes the latest automatic control systems in continuous sheet and tinplate mills and shows how their introduction has led to better quality materials.—M. D. J. B.

All-Union Conference of Tube-Industry Workers. N. F. Ermolaev. (*Stal'*, 1956, (10), 945–947). [In Russian]. 260 delegates attended the tube-industry conference at Pervoural'sk, 4–9th July, 1956. Among the points discussed were the comparatively small relevance of research work to actual tube industry problems, the failure of the industry to produce the needed type of tube and the future development of the industry.—S. K.

Production of Thick-walled Large and Medium Diameter Seamless Tubes. S. I. Borisov. (*Stal'*, 1956, (10), 905–908). [In Russian]. The Soviet oil industry is the main consumer of the tubes (145–630 mm dia.) whose production in Pilger and automatic mills is compared in this article. The tube-quality requirements of the oil industry are considered and it is concluded that several technological and design modifications of both types of mill must be carried out.—S. K.

A Statistical Control Method for Tube Production. I. M. Ludenskii. (*Stal'*, 1957, (6), 543–547). Correlation between average wall thickness and thickness variations led to a statistical method of dimensional control. Rolling practice has been considerably improved.

Automation of the Operation of Tube Rolling Mills. M. E. Katsnel'son. (*Stal'*, 1956, (11), 1000–1005). [In Russian]. Since 1949 seven tube mills in the U.S.S.R. have been automated. In this article the considerable experience accumulated at the following of these is analysed: the Lenin at Dnepropetrovsk, the Zakavkazskii Metallurgical, the Bakinskii Tube and the Pervoural'skii Novotrubnyi. It is concluded that further improvements in the tube-rolling industry are impossible without full mechanization and automation, and the measures required for securing complex automation of the whole hot-rolling process from the charging of the billet into the continuous furnace to the delivery of the finished tube for inspection are listed. Automation equipment is surveyed.—S. K.

Improving the Kinematics of Mills for Cold-Rolling Tubes. I. Yu. Korobochkin. (*Stal'*, 1956, (9), 806–808). [In Russian]. Relatively simple changes in the construction and operation of Soviet-operated American tube cold-rolling mills which have improved their operation and enabled roll design to be standardized are briefly described.—S. K.

Wall Thickness Variations of Tubes when Rolled in a Three-Roll Mill. P. K. Teterin, F. A. Danilov, and E. S. Trifonov. (*Stal'*, 1956, (8), 721–727). [In Russian]. An account is given of a detailed, full-scale investigation of wall-thickness variations in tubes rolled from pierced billets in a three-roll and also in an automatic mill. Factors studied included the actual thickness difference with normal lengths of end-cutting (not only the maximal wall-thickness deviation but also the components of this deviation), transverse wall-thickness variation for different lengths of end cutting and changes in the original wall-thickness variation of the pierced billet when rolled. The three-roll mill was found to be superior to the automatic.—S. K.

Application of Eddy-Current Adjustable Drives in the Steel Industry. T. Skarakis. (*Iron Steel Eng.*, 1957, **34**, May, 95–98). The author describes the basic principles of operation of the eddy-current coupling used in conjunction with squirrel cage or synchronous motors to provide a wide range of stepless, adjustable speeds from an a.c. power source. The efficiency, cooling and control of the unit are discussed, as well as its application to mill operation.—M. D. J. B.

MACHINERY FOR IRON AND STEEL PLANT

Rational Operation of Pit Cranes. P. Z. Petukhov and V. N. Sutorikhin. (*Stal'*, 1957, (6), 550–552).

Conversion of Electric Bridge Cranes to Rolling Friction Bearings. V. Ya. Grigor'ev. (*Stal'*, 1956, (10), 941–942). [In Russian]. The conversion of soaking-pit cranes at the A. K. Serov works from sliding to rolling friction bearings is briefly described. The conversion is said to give an annual saving of about 8000 roubles per crane.—S. K.

Better Accounting For Rail Transport Inside Works. V. S. Perlin. (*Stal'*, 1957, (1), 72–75). [In Russian]. The author criticizes present methods used in Soviet iron and steel works for intra-works rail transport accounting, especially the t-km unit. He urges the adoption of a generally agreed, uniform and improved system in which account is to be taken of railway-shop operating data.—S. K.

Electrification of Rail Transport in Iron and Steel Works. E. A. Khokhlov. (*Stal'*, 1956, (9), 835–841). [In Russian]. Existing types of Soviet diesel-electric and electric locomotives are considered unsuitable for the fullest use within iron and steel works. The locomotive characteristics required for the complete electrification of intra-works rail transport under various conditions are formulated.—S. K.

Direct-Current High Potential Insulation Testing. L. Shaw. (*Iron Steel Eng.*, 1957, **34**, May, 119–122). The author proposes a method of high voltage testing for evaluating the condition of motor and generator insulation.

Productive Maintenance. R. G. Rettig. (*Iron Steel Eng.*, 1957, **34**, May, 117–119). The author discusses the advantages of planned maintenance and shows how this can lead to planning machine overhauls well in advance of costly production stoppages.—M. D. J. B.

WELDING AND FLAME CUTTING

On Seam-Welding of Mild Steels. M. Lipa. (*Zváranie*, 1956, **5**, (3), 79–82). [In Slovak]. Details are given of a study of two low-C steels which, contrary to expectation, showed weld-brittleness.—P. F.

Resistance Welding of High-Speed Steels. B. Vrána. (*Zváranie*, 1955, **4**, (8), 241–250). [In Czech]. A detailed discussion is given of the technology of welding and heat-treating high-speed steel tool-tips.—P. F.

Repairs on Grey Cast Iron Parts by Electric Arc Welding. B. Zorkóczy. (*Zváranie*, 1955, **4**, (2), 37–40). [In Slovak].

Details are given of the development and method of application of a composite consumable electrode for welding grey cast iron.—P. F.

Welding Pre-Heated Large Grey Cast-Iron Pieces, Propane-Butane Burners Being Used for Pre-Heating, R. Halbich. (*Zváranie*, 1956, **5**, (3), 87–90). [In Czech].—P. F.

Important Procedures in Proper Welding Shop Management. J. L. Fortunato. (*Welding J.*, 1957, **36**, June, 567–575). Planning, identification of materials, techniques, particularly automatic methods and testing methods are discussed.

Welding of Castings made from Stainless Steels of the 24/12 Cr–Ni Type. J. Škriniar. (*Zváranie*, 1956, **5**, (5), 150–155). [In Slovak]. The technology of welding and weld-testing is discussed in detail.—P. F.

Welding or Casting? B. Wetterström and E. Gladh. (*Gjuteriet*, 1957, **47**, Mar., 37–41). [In Swedish]. The relative economic merits of welding and casting are discussed with reference to the 30 years' old design of a 68-tons steam turbine casing. The old design can be modified to make it suitable for welding, in which form it is cheaper to produce than a casting. On the other hand, materials and casting methods have been improved, so that a suitably modified design may enable casting to compete with welding.—L. D. H.

Automatic Welding Up of Rolls. V. K. Petrichenko. (*Avtomaticheskaya Svarka*, 1955, (6), 74–84). [In Russian]. The importance of automatic methods of welding up rolls is emphasized. Many tabulated data on the economics of the process are presented. These include the factors of steel quality, period of operations of the rolls between repairs, number of welding up repairs until complete failure of the roll, costs, etc. An automatic roll welding device is described.

The Theory of Pressure Welding. F. Erdmann-Jesnitzer. (*Zváranie*, 1956, **5**, (8–9), 228–233). [In Czech]. The theoretical aspects of pressure welding are surveyed, with special reference to the author's own researches on metals and ionic salts.—P. F.

Automatic Welding Under a Ceramic Flux of Parts Subject to Abrasion. M. S. Samotryasov. (*Avtomaticheskaya Svarka*, 1957, (1), 40–45). [In Russian]. High-chromium welds laid down under a ceramic flux exhibit good resistance to abrasion and have excellent impact strength. The weld metal contains 2·5 to 3% C, 26 to 30% Cr, 2 to 2·8% Mn, 1·2 to 1·5% Si. As distinct from the Sormite type of metal, this weld metal does not contain Ni which coarsens the primary structure and intensifies the dendritic heterogeneity of the weld. The absence of nickel has practically no effect on resistance to abrasion.—R. S.

Improvement in the Design of the Pneumatic Mechanism of Spot Welding Machines. D. S. Balkovets. (*Avtomaticheskaya Svarka*, 1954, (2), 56–65). [In Russian]. B. presents a theoretical analysis of the operation of the pneumatic mechanism of a spot welding machine. It is shown that, in existing types of machine, the pressure of the electrodes during welding varies due to the delay in filling the cylinder with air during high speed operation. A system is described which ensures the constancy of electrode pressure during welding.—R. S.

Welding of Niobium-Stabilised Stainless Steels for use in Turbines at 600° C. J. Němec. (*Zváranie*, 1956, **5**, (1), 18–21). [In Czech]. The author discusses Curran and Rankine's article on welding type 347 stainless steel for 1100° F, which appeared in the *Welding Journal* in March, 1955.

New Fusion Methods Ease Joining of Hard-to-Weld Materials. C. B. Robinson. (*Western Metals*, 1957, **15**, May, 54–56). Inert gas-shielded cutting processes, metal inert-gas and tungsten inert-gas welding, and the use of iron powder electrodes are outlined.

Removing Weld-Stresses by Flame Annealing. H. G. Kunz. (*Zváranie*, 1956, **5**, (3), 73–79). [In Czech]. Practical methods are reviewed. The primary source of the article is a contribution on the above subject which appeared in "*Schweissen und Schneiden*" in July, 1955.—P. F.

Weldability of Alloy Steels by the Submerged-Arc Method. Z. Bajer. (*Zváranie*, 1956, **5**, (10), 311–315). [In Czech]. The weldability of low- and medium-, as well as high-alloy steels, including stainless and other acid-resisting types by this method is discussed. Special reference is made to relevant fluxes and electrodes, mainly as used in the U.S.S.R. The examples shown indicate that the complex problem of welding high-alloy steels can be solved satisfactorily.—P. F.

Welding Kovar with Steel Electrodes. A. Kloss. (*Zváranie*, 1956, **5**, (11), 330–331). [In Czech]. The possibility of welding Kovar, an alloy containing, in addition to iron, 28·5% Ni and 18% Co, by means of low-carbon electrodes was investigated. The results showed the method to be inapplicable without further studies of the Fe–Co–Ni system. Kovar tubes welded together with steel electrodes were prone to crack-formation due, it is assumed, mainly to differences in the thermal expansion of the alloy and the welds.—P. F.

Automatic Argon-Arc Welding of Special Steels. J. Višňovec. (*Zváranie*, 1956, **5**, (12), 365–372). [In Czech]. Details are given of the welding technology recommended with automatic equipment of Soviet and Czechoslovak manufacture in welding 18/8, 24/19 and 16/19 type stainless steels. The recommendations are based on the author's researches, which are discussed.—P. F.

Programme Control of the Process of Contact Spot Welding. Yu. A. Pachentsev. (*Avtomaticheskaya Svarka*, 1957, (1), 77–79). [In Russian]. A scheme for the programme control of the process of spot welding is described, based on utilizing the effect of the movement of the electrodes of the welding machine.—R. S.

Brazed Stainless Sandwich Spotwelded. Herr, R. C. Smith and G. L. Peterman. (*Metalworking Prod.*, 1957, **101**, Mar. 29, 547–549). Corrosion tests are included and were satisfactory.

Forging of Welds—A Method for Controlling Weld-Stresses and Preventing Cracking. J. Nemec. (*Zváranie*, 1955, **4**, (2), 41–43). [In Czech]. The scope of weld-forging, details of the method, and the testing of such welds are discussed with special reference to electric-arc welding of mild steel.

Welding of Austenitic Acid-Resisting EI–530 and EI–629 (CrNiMoCu) Titanium-Stabilised and EI–628 (CrNiMo) Titanium-Stabilised Steels. A. S. Astaf'ev and A. V. Russiyan. (*Avtomaticheskaya Svarka*, 1956, (3), 72–78). [In Russian]. The compositions of the steels were as follows: EI–530: 0·06 C; 0·56 Si; 0·61 Mn; 0·008 S; 0·0018 P; 17·45 Cr; 29·60 Ni; 2·80 Mo; 0·21 Ti; 3·95 Cu. EI–629: 0·04 C; 0·38 Si; 0·50 Mn; 0·008 S; 0·0250 P; 23·15 Cr; 27·40 Ni; 3·50 Mo; 0·34 Ti; 3·36 Cu. EI–628: 0·04 C; 0·32 Si; 0·52 Mn; 0·010 S; 0·0230 P; 23·20 Cr; 26·82 Ni; 2·46 Mo; 0·40 Ti;—Cu. A method is described for the manual arc welding and automatic welding with flux-coated electrodes. The tests were carried out on 2 mm thick plates at 152–162 feet per hour. As regards resistance to 20% boiling sulphuric acid, the welds of EI–530 and EI–629 steels are not inferior to the base metal. The mechanical properties of EI–629 and EI–628 were similar to those of the base metal but EI–530 was prone to intergranular corrosion.—R. S.

Investigations of Some Factors Which Determine Arc Stability Under a Flux. G. I. Leskov. (*Avtomaticheskaya Svarka*, 1956, (3), 17–25). [In Russian]. An experimental study was made of the arc under a flux after cutting off the current. After the current has passed through zero between the electrodes, the liquid flux seal is preserved and plays an important part in the repeated excitation of the arc in each semi-period. To improve a.c. arc stability under a flux, fluxes must be developed with physico-chemical properties which will ensure the presence of liquid seals in each half period, their rapid volatilization by the current and good conditions for development of the arc after volatilization. Such properties include high viscosity throughout a wide range of temperature, maximum electrical conductivity associated with maximum development of energy, low specific heat, low boiling point (which apparently is in contradiction to the high viscosity previously mentioned) and low heat of evaporation of the flux.—R. S.

Local Heat Treatment of Welded Joints. B. S. Kasatkin and A. A. Rossoshinkii. (*Avtomaticheskaya Svarka*, 1957, (1), 31–36). [In Russian]. An irregular distribution of second and third magnitude stresses was observed in low alloy steel welds and this was associated with a block structure. Local heat treatment not only refined the secondary structure of the surface layers of the weld, but also modified the fine structure and led to a marked reduction in stresses. The impact strength of the weld was improved.—R. S.

Oxygen-Sand Cutting of Cr–Ni Acid-Resisting Steels and Clad Steels. G. A. Ukolov, V. N. Chernov and N. N. Korf. (*Avtomaticheskaya Svarka*, 1956, (2), 89–92). [In Russian]. The oxygen-sand cutting device is illustrated diagrammatically. It ensures stability of cutting, is cheap (sand is

used instead of electrodes and iron powder) and it is more productive than arc and oxy-flux cutting. The method facilitates the cutting of stainless steel plate 4 to 100 mm thick and clad steel up to 46 mm thick by reducing tolerances and leaving a smooth surface after cutting.—R. S.

MACHINING AND MACHINABILITY

Machining Properties of Iron Castings. N. Zlatin, W. H. Friedlander, and C. F. Walton. (*Metalworking Prod.*, 1957, **101,** July 12, 1201–1208). **General Machining Recommendations.** (July 19, 1245–1249). An extract from a forthcoming "Gray Iron Handbook." Microstructure, especially graphite structure, is reviewed and machinability data, surface finish and coolants, turning operations, power requirements, milling, drilling and other processes and a section on economics are included, with many curves and tables.

Metallurgical Control in Cutting Tool Manufacture. J. G. Ritchie. (*Australasian Eng.*, 1957, **49,** Apr. 8, 62–70). A review of the materials used, their properties and heat treatment is given. Quality is considered and examples of defects and testing methods are briefly discussed.

The Machinability Concept. K. G. Lewis. (*Metal Treatment and Drop Forging*, 1957, **24,** July, 263–271). A review of methods of estimation is given with mathematical treatment of the various parameters suggested. (42 references).

High Rake Cutters for High Tensile Steels. P. Trippe. (*Metalworking Prod.*, 1956, **100,** Oct. 26, 1723–1729). Designs and tests on slab milling cutters are described, also on side-and-face cutters. Some experimental cutters are also mentioned. It is hoped to carry developments into the 70–75 t range.

The Influence of Lead on Metal-Cutting Forces and Temperatures. M. C. Shaw, P. A. Smith, E. G. Loewen, and N. H. Cook. (*Trans. Amer. Soc. Mech. Eng.*, 1957, **79,** July, 1143–1154). The effects of adding S and Pb to steel on tool-face friction, cutting energy per unit volume and tool-face temperature were studied by high-speed photography. Interference due to some cutting fluids occurred. Tool life cannot be predicted from the above parameters.

Leaded Steel and the Real Area of Contact in Metal Cutting. M. C. Shaw, P. A. Smith, N. H. Cook, and E. G. Loewan. (*Trans. Amer. Soc. Mech. Eng.*, 1957, **79,** July, 1165–1171). The action of lead is discussed and the thickness of the layer of lead effecting lubrication is found to be small. Measurements on a steel with and without Pb enable real arc of contact to be calculated as 1–2% of apparent area. Presence of lead makes cutting fluids such as CCl_4 less sensitive to increased speed.

Cutting Fluid Evaluation. H. W. Husa. (*Trans. Amer. Soc. Mech. Eng.*, 1957, **79,** July, 1172–1176). A tracer element study of the transfer of tool material to chip was made. 1015 and 1045 steels were used and results for typical compound and mineral oils were obtained.

Acid Etch-Milling Is In Full-Scale Production. T. M. Rohan. (*Iron Age*, 1957, **179,** May 16, 114–115). Applications of etch-milling of Al are described. Development of the process for use on Ti and steels is reported.—D. L. C. P.

CLEANING AND PICKLING

The Cleaning and De-Rusting of Steel. K. A. van Oeteren-Panhäuser. (*Metalloberfläche*, 1957, **11,** Feb., 60–63; Mar., 89–97). Methods for cleaning and removing rust from steel as a preliminary treatment before applying protective coatings are reviewed. Examples from practice are given and equipment described. (24 ref.).—L. D. H.

Basic Data on Mechanical Cleaning Operations in Steel Plants. G. D. Dill. (*Iron Steel Eng.*, 1957, **34,** May, 123–132). Modern methods of mechanical cleaning and descaling are described. Application to the cleaning of billets, strip, sheet and plate, wire rod and structural steel is discussed.

Electrolytic Polishing of Metals. I. Hrivňák. (*Hutnické Listy*, 1956, **11,** (10), 599–611). [In Slovak]. Recent developments in the theory and practice of electrolytic polishing of metals are surveyed. Details are given of electrolytic polishing machines and mixers for the safe mixing of $HClO_4$ and organic acids as developed in the Czechoslovak Welding Institute.—P. F.

Removing Scale by Pickling. F. Matzinger. (*Hutník*, 1957, **7,** (1), 22–24). [In Slovak]. Experiments on the solubility of various oxides of iron in acids are reported. It is shown that by influencing oxidation processes by means of process variables such as annealing temperatures, protective or partly protective atmospheres etc. it is possible to favour the formation of one type of scale rather than some other, i.e. one more easily removable by pickling.—G. F.

Advances in Pickling Technology. M. Jaroš. (*Hutník*, 1956, **6,** (11), 335–336). [In Czech]. The present state of knowledge relating to pickling is outlined, and methods in practical use are considered. The use of inhibitors, gaseous pickling methods, spray-pickling, regeneration of waste pickling liquors and other topics are surveyed.—P. F.

Pickling Flaws in Steel due to Hydrogen. A. Dvořák. (*Hutník*, 1956, **6,** (11), 337–341). [In Czech]. The origin and prevention of flaws and embrittlement are discussed.

A New Pickling Inhibitor. M. Jaroš. (*Hutník*, 1957, **7,** (1), 24–25). [In Czech]. The effectiveness of a commercially available inhibitor known as DBS, developed by the East Bohemian Chemical Works is described. Steels pickled for 30 min in dilute H_2SO_4 or HCl containing the inhibitor showed no measurable increases in hydrogen content while, if pickled under similar conditions without inhibitor, appreciable hydrogen uptake was observed to take place.—P. F.

PROTECTIVE COATINGS

The Determination of the Adhesive Power of Metallic Coatings on Wires. E. A. Franke. (*Metalloberfläche*, 1957, **11,** Feb., 53–59). It is considered that existing tests are not completely satisfactory in ascertaining the ability of the metal coating to resist various types of stressing, and a number of tests have been developed by the author, and are here described.—L. D. H.

What Size Filter? G. Gabrielson. (*Metal Finishing*, 1957, **55,** July, 50–51). An equation is derived for continuous filtration (of electroplating solutions).

Ion Exchanges in Electroplating Technique. G. Gabrielson. (*Metalloberfläche*, 1957, **11,** Feb., 41–46). The fundamental properties of the most important ion exchange resins are described, and their application to the analysis of electroplating solutions is discussed. (25 references).—L. D. H.

Hard Coatings and Surfaces for Metals. R. J. Fabian. (*Mat. Methods*, 1957, **45,** Jan., 121–140). The processes of hard facing, diffusion coating electrodeposition, anodic coating and spray coating for increasing surface hardness are described and compared in detail. Typical uses of each process are given.—A. D. H.

Chromium Plate Irregular Surfaces Uniformly. L. Rosenberg. (*Iron Age*, 1956, **178,** Nov. 15, 144–146). Interior plating of cannon is described. Anodes closely adjacent to the surface are used with high amperage and a hot bath. Films up to 0·007 in. thick free from porosity can be produced.

Chromizing for Resistance to Corrosion and Wear. D. E. Lehane and R. L. Wachtell. (*Product Eng.*, 1956, **27,** Nov., 180–183). The production of either ductile or hard cases by gaseous diffusion is outlined and typical uses are tabulated.

High Temperature Corrosion Resistance of Chromized Coatings Applied to Steel by the Gaseous Method. (*Canad. Mines Techn. Surveys, Research and Special Projects (July-Dec.* 1956), 1957, Feb. 8, 22–23).

Bright Nickel Plating on Cast Iron. E. Alary. (*Electroplating Metal Finishing*, 1957, **10,** July, 227, 230). A sequence of ten operations, (degreasing, rinsing, etc. and alkaline copper plating) followed by Ni and Cr plating gave satisfactory finishes.

Influence of Heat Treatment on the Properties of Chemical Nickel Coatings Produced by the Kanigen Process. V. Royen. (*Electroplating*, 1957, **10,** April, 114–115). The results of chemical, hardness and metallographic tests show that annealing at 400° C gives a hardness of 1200 V.P.N., annealing at 600° C gives a compromise between maximum hardness and maximum ductility. Annealing at 750° C followed by slow cooling to 200° C gives the best corrosion resistance.

Ion Exchange in a Modern Nickel-Chromium Plating Plant. (*Electroplating*, 1957, **10,** April, 116–118). The plant and processes used in an ion exchange treatment method for the recovery from plating wastes of chromium, copper and nickel are described.—A. D. H.

The Use of Fluxes in Hot-Galvanizing. K. Janecký. (*Hutník*, 1956, **6,** (10), 301–304). [In Czech]. The function and efficiency of fluxes are discussed. It is shown that the use of dry flux, i.e. zinc chloride from which the water is removed by evaporation after application to the metal, is preferable to

the "wet" method in which the steel sheet is not passed through ovens to effect removal of the water from the flux.

Effects of Impurities and Purification of Electroplating Solutions. I. Nickel Solutions. 9, The Effect and Removal of Aluminium. D. T. Ewing, A. J. Smith, and W. O. Dow. (*Proc. Ann. Conv. Amer. Electroplaters' Soc.*, 1956, **43,** 44–46). **10, The Effect and Removal of Manganese.** A. J. Smith and R. J. Rowe (46–49). Al has little effect on coatings. Precipitation is complete at pH values over 5 and only 10 mg/l remain at pH 4·5. Manganese has little effect, and is difficult to remove.

Mass Spectrometric Examination of Hydrogen in Chromium-Plated Steel. C. Levy and G. A. Consolazio. (*J. Electrochem. Soc.*, 1956, **103,** Nov., 624–627). The evolution of hydrogen from chromium plated mild steel in the range 0–800° C was studied using the mass spectrometer. Above 550° C Cr plating had no effect on the rate of evolution. Below 350° C, H_2 is evolved only from the plating while above 490° C the evolution occurs mainly from the basis steel. No specimen could be completely degassed even after heating for 20 h at 800° C. Most of the hydrogen from plated and unplated specimens was evolved at 350° C after 80 min heating.

Continuous Strip Galvanizing at Inland Steel. A. Q. Smith. (*Products Finishing*, 1957, **21,** Jan., 68–74). The plant and methods at the Indiana Harbor Works of Inland steel for the production of galvanized strip by the Sendzimir process is described with reference to the controlled atmosphere annealing furnaces.—A. D. H.

Get More Protection from Galvanized Coatings. A. T. Baldwin and W. H. McMullen. (*Iron Age*, 1957, **179,** Mar. 28, 115–117). A review of the adhesion of thick zinc coatings to steel and metallographic studies on cracking and the formation of Fe–Zn alloy layers. The use of Al is noted.

Method of Automatic Control of a Continuous Heurtey-Sendzimir Galvanizing Line. A. Doyle. (*Silicates Indust.*, 1957, **22,** Feb., 87–94). A brief illustrated account of the furnaces and other features of the Sendzimir installations, especially that of S. A. Phenix. Speed, tension, guides, temperature regulation both in gas-heated and electrically heated zones and advantages of the system are briefly considered.

The Behaviour of Copper in Steel During Scaling and Its Effect on Galvanizing Steel Wire. C. Eisenhuth, W. Gruhl, W. Papsdorf, and I. Eisenhuth. (*Stahl u. Eisen*, 1957, **77,** March 21, 354–359). The Cu (0·38%) in steel segregates at the grain boundaries of the surface layer after annealing in an oxidizing atmosphere, but there is no segregation in neutral-atmosphere annealing e.g. in argon. The metallic Cu segregated at the grain boundaries is heavily attacked by the molten Zn of the galvanizing bath, frequently resulting in cracking of the wire.—T. G.

The Diffusion of Tinplate Manufacture. W. E. Minchinton (*Economic History Review*, **9,** (2), reprint 349–358). Historical.

Tinplate Manufacture. Productivity at Recently Installed Plants in South Wales. J. H. Mort. (*Iron Steel*, 1957, **30,** Apr., 131–136; May, 183–185). The equipment is briefly described (cold-reduction mills and electro-tinning lines), and operational formulæ in terms of weight, thickness and other factors per base box are derived and applied. Graphical solutions are then shown and nomograms presented. The improvements in output secured by the more modern lines are indicated. (10 references).

Flow-Brightening of Tinned Ware. C. J. Thwaites and W. E. Hoar. (*Tin and its Uses*, 1957, Summer, 15). A note on the process.

Surface Characteristics of Tinplate. W. E. Hoare. (*Tin and its Uses*, 1957, Summer, 9–11). Natural and artificial films, including oil films, are reviewed.

Aluminizing. (A Review of the Literature up to 1955). D. Horstmann. (*Stahl u. Eisen*, 1957, **77,** March 21, 359–362). An illustrated review dealing with pretreatment, e.g. pickling, use of protective coatings, and processes using fluxes or protective atmospheres. (110 references).—T. G.

Aluminized Steel Stands Off Atmospheric Corrosion. J. C. Merritt and W. E. McFee. (*Iron Age*, 1956, **178,** Dec. 27, 60–61). An Al-coated sheet is described with properties, temperature resistance, uses and corrosion properties.

Applying Aluminium to Prevent Iron from Scaling and Corrosion. U. Hollo. (*Kohászati Lapok*, 1956, **11,** (10), 220–223). After referring to the various kinds of aluminium coating and their application, the methods used in Hungary are particularly discussed: (1) The *spraying method* or *Alumetier process*, when the metal spraying is followed by heat treatment and finally a protective coating is applied. (2) *Alitizing*, a process of heating iron or steel in finely powdered mixture of aluminium and aluminium oxide or aluminium–iron alloy and aluminium oxide or bauxite according to a new method developed in Hungarian Iron and Steelwork Diósgyör. In all these methods ammonium chloride is used as catalyst. The alitized iron and steel is reputedly heat resisting up to 900° C. (3) Aluminium coating by plunging in molten aluminium, i.e. *hot dip aluminizing*.—P. K.

"Cold Galvanising," a Novel and Universal Protection Against Corrosion. H. Edel. (*Maschinenwelt Elektrotechnik*, 1956, **11,** Aug., 251). An outline is given of the properties and range of application of zinc-rich paints.—L. D. H.

Control of Corrosion with Zinc Coatings. J. L. Kimberley. (*Corrosion*, 1957, **13,** June, 385–391). The theoretical background and an account of the development of the industry are given in brief and descriptions of hot-dip, spray, sherardizing, and painting. The fields of usefulness of the various coatings are outlined and case histories are cited.

Improving the Technology of Hot Tinning. P. G. Kovtun. (*Stal'*, 1956, (10), 938–940). [In Russian]. Experiments on the pre-tinning treatment of steel, the hot tinning and the removal of the excess of tin in oil baths are described, most attention being given to the last. By adopting the recommended procedures the technology of the hot-tinning process can, it is claimed, be improved.—S. K.

Aluminizing of Cast Iron Ingot Mould. N. P. Nikolaichik and E. N. Nikolaichik. (*Stal'*, 1957, (6), 571–573). Calorizing by mixed powders, in baths of molten Al and by metallization and diffusion annealing were used, and the structures of the layers formed are described. The life of the moulds was more than doubled, 6-t moulds after coating serving for 115–127 casts when longitudinal cracking occurred, but crazing was practically eliminated.

Study on Aluminium-coated steel. I. On the Oxidation of Aluminium-coated Steel Sheet at High Temperature. T. Saga and O. Miyakawa. (*Nippon Kinzoku Gakkai-Si*, 1955, **19,** July, 404–408). [In Japanese]. Hot-dipped aluminized sheet was tested at 800–1000° C. The increase in weight from oxidation was reduced by thicker alloy layers, thinner amounts of base metal, and higher carbon contents in the base metal. The amount of excess Al layer had little effect on heat-resistance. Rapid heating and cooling caused a thin alloy layer to flake.—K. E. J.

Measuring the Thickness of Sprayed Metal Coatings. R. E. Mansford. (*Electroplating Metal Finishing*, 1957, **10,** July, 208–212, 220). A comparison of commercially available instruments is described. Moving-iron and electromagnetic instruments are better than pull-off gauges, but calibration curves can be plotted for the latter by the operator. Characteristics and limitations of the various types and the best conditions for use are discussed.

Passivation of Iron by Chromate Solutions, D. M. Brasher, A. H. Kingsbury, and A. D. Mercer. (*Nature*, 1957, **180,** July 6, 27–28). D. M. Brasher and C. P. De (28–29). Uptake of Cr is plotted as a function of time of immersion for various times of pre-exposure to air. Interpretation is suggested in terms of growth mechanism. The influence of O_2 in the solution is important and competition with CrO_4-ion apparently occurs. A picture of inhibition by chromate is then sketched.

Jet Lapping as a Preliminary Treatment for the Phosphating of Sheet. E. Klüppelberg. (*Metalloberfläche*, 1957 **11,** Jan., 17–18). Comparative tests showed that an adherent coating of lacquer can best be produced on a surface prepared by jet lapping, followed by phosphating.—L. D. H.

Coal Tar Coatings for Protection of Underground Structures. N. T. Shideler. (*Corrosion*, 1957, **13,** June, 392–396). Specifications for hot-applied coatings and optimum temperature ranges in service are given. A thixotropic cold-applied material is also described.

Phenolic Coatings for Metal Products. R. B. Young. (*Mat. Methods*, 1957, **45,** Jan., 106–109). The properties, application, uses and costs of phenolic plastic coatings are discussed.—A. D. H.

A New Process for the Production of Plastic-Coated Sheet: "Mechanical Bond" Eliminates Adhesives. (*Sheet Metal Ind.*, 1957, **34,** July, 526–527). The sheet is perforated and the surrounding areas dished to obtain a keying surface.

Cold Dip Strippable Compounds. M. H. Sandler. (*PB* 118321, 1952, *TIDU List*, 742, pp. 23). Cellulose acetate-butyrate with plasticisers, resins, and anticorrosives formed a mobile solution at 75–100° F. All inhibitors tried protected Al, some also Cu and others also steel, none protected all three metals.

Porcelain Enamels: What the Properties Mean. J. H. Giles, jun. (*Mat. Methods*, 1956, **44,** Oct., 108–111). Standard chemical, mechanical and optical tests for porcelain enamel are described and the significance of each is discussed.

Development of Enamels. W. G. Ball. (*Inst. Vit. Enamel.*, 1957, **7,** July, 83–86). A general account of frit composition and uses of enamel coatings is given.

Surface and Subsurface Abrasion Tests for Porcelain Enamels. J. H. Giles jun., and J. C. Richmond. (*Amer. Ceram. Soc. Bull.*, 1957, **36,** June, 205–207). Apparatus is described in which abrasive, stainless steel balls and water are used and loss of gloss is taken as an index of surface and loss of weight as an index of subsurface abrasion. Enamels were found to rank differently for the two measures of resistance.

Selection of a Protective Coating for Steel against Corrosion. (*Acier*, 1957, pp. 36). An illustrated review of the painting of structural steel. The preparation of the surface and discussion of the types of coating available with their properties, and their selection in view of the conditions of the environment are presented.

POWDER METALLURGY

Designing for Powder Metallurgy Fabrication. (*Precision Met. Mold.*, 1957, **15,** Apr., 72–74, 76–78). Various practical points are indicated with a check-list, and coating is considered in some detail.

Which Metal Powder Part? R. Talmage, W. J. Dollker, and C. G. Johnson. (*Mat. Methods*, 1956, **44,** Oct., 116–117). The physical and mechanical properties of medium density, high density, and infiltrated parts made by powder metallurgy are compared and typical applications of each are mentioned.—A. D. H.

The Effects of Initial and Subsequent Pressure and of the Sintering Temperature on the Properties of Sintered Components Made of Powdered Iron. G. Bockstiegel. (*Arch. Eisenhüttenwesen*, 1957, **28,** Mar., 167–174). The dependence on pressures and sintering conditions of specific gravity, tensile strength, breaking elongation, electrical conductivity, and compressibility has been investigated in sintered specimens of 4 different iron powders produced by reduction, electrolytic methods or mechanical comminution.—L. J. L.

Metal Powders. Commercially Available Grades and Types. (*Metal Powder Association Data Sheet*, No. 3 Revised Feb. 1956, pp. 6).

Method of Sieve Analysis of Granular Metal Powders. (*Metal Powder Association Standard*, 5–46. Approved 1949, pp. 3).

Subsieve Analysis of Granular Metal Powders by Air Classification. (*Metal Powder Association Standard*, 12-51T. Issued Jan. 1951, pp. 4).

Determination of Iron Content of Iron Powder. (*Metal Powder Association Standard*, 7–54T. Revised 1954, pp. 6).

Determination of Insoluble Matter in Iron and Copper Powders. (*Metal Powder Association Standard*, 6–54T, Revised 1954, pp. 4)

Determination of Green Strength of Compacted Metal Powder Specimens. (*Metal Powder Association Standard*, 15–51. Adopted 1955, pp. 5).

Determination of Bending Strength, Green Density, Hardness, and Shrinkage of Compacted Sintered Metal Powder Specimens. (*Metal Powder Association Standard*, 13–51T. Issued 1951, pp. 4).

Standards, Specifications and References for Metal Powders and Metal Powder Products. (*Metal Powder Association Data Sheet*, No. 4. Revised Feb. 1956, pp. 4).

Typical Properties of Powder Compacts. (*Precision Met. Mold.*, 1957, **15,** Apr., 70–71). Compositions, strengths and hardnesses are tabulated.

Electronic Core Materials and Core Manufacturer's Designations. (*Metal Powder Association Data Sheet*, No. 1. Revised Feb., 1956, pp. 2).

Metal Powder Sintered Bearings (Oil Impregnated). (*Metal Powder Association Standard*, 16–55T. Revised Nov., 1955, pp. 4).

Metal Powder Sintered Bearings (Oil Impregnated) Preferred Dimensional Specifications. (*Metal Powder Association Standard* 14–51T. Issued June, 1951, pp. 2).

Electronic Iron Core Preferred Dimensional Specifications. (*Metal Powder Association Standard*, 11–57. Issued Jan., 1957, pp. 4).

Tension Test Specimens for Pressed and Sintered Metal Powders. (*Metal Powder Association Standard*, 10–50. Adopted 1953, pp. 3).

Iron Powder Rings Replace Cold Rolled Steel. (*Precision Met. Mold.*, 1957, **15,** June, 44–45). Magnetic-return-path rings for electric meters are fabricated from powder.

FERRITES, CERMETS AND CARBIDES

Comparison of the Strength of Sintered Carbides. J. Gurland. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209;** *J. Met.*, 1957, **9,** Apr., Section 2, 512–513). The author has determined and compared the transverse rupture strengths of several carbide-cobalt sintered compacts. The rupture strength appears to be related to the elastic modulus of the carbide used.—G. F.

Improved Silicon Carbide for High Temperature Parts. K. M. Taylor. (*Mat. Methods*, 1956, **44,** Oct., 92–95). Physical and mechanical properties of "KT" SiC which is self-bonded are superior to silicate–silicon nitride or silicon–bonded material. It is stronger, has a lower permeability, better thermal properties and higher abrasion resistance.

Magnetic and Magnetostrictive Properties of Magnesium–Nickel Ferrites. P. O. Hoffmann. (*J. Amer. Ceram. Soc.*, 1957, **40,** July, 250–252).

Dimensional Changes in Nickel Zinc Ferrite in the Neighbourhood of the Curie Temperature. R. C. Deshpande. (*Trans. Indian Inst. Met.*, 1954–55, **8,** 253–259).

PROPERTIES AND TESTS

The Strength of Materials. J. Němec. (*Hutnické Listy*, 1956, **9,** (10), 583–591). [In Czech]. The theoretical and practical aspects of ductile, brittle and fatigue fractures are discussed. Heterogeneous stress distributions, surface and size effects are considered.—P. F.

Development of the Micro-Mechanical Method of Testing Materials. I. M. Roĭtman. (*Zavodskaya Laboratoriya*, 1956, **22,** (7), 834–840). [In Russian]. Micro-mechanical testing machines are divided into two groups according to the principles of their operations. Examples of each type are considered, their construction and operating characteristics being outlined.—S. K.

Graphite Resistor High Temperature Furnace. (*Metallurgia*, 1957, **55,** May, 255–256). A brief description is given of a furnace designed to give a temperature of 2000° C and constructed by G.E.C. Ltd.—B. G. B.

Evaluating the Quality of Steel Sheet from the Fracture Structure. P. M. Dontsov. (*Stal'*, 1956, (9), 842–843). [In Russian]. The mechanism of the formation of fracture structures in steel-sheet test pieces and the possible ambiguities in quality evaluations based on these structures are discussed. The two possible types of lamination, that existing in the metal before testing and that developed during the test, must be distinguished.—S. K.

Further Studies of the Hot-Ductility of High-Temperature Alloys. E. F. Nippes, W. F. Savage, and G. Grotke. (*Welding Res. Council Bull. Series No.* 33, 1957, Feb.). The effect of weld thermal cycles on the elevated temperature properties of 34 different materials has been studied. The results of the hot-ductility test show marked differences between the elevated temperature properties of different alloys.—U. E.

On the Ductility of Iron at 4·2° K. Z. S. Basinski and A. Sleeswyk. (*Acta Met.*, 1957, **5,** March, 176–179). A letter. Tensile tests on vacuum-refined iron at 4·2° K show that ductile fracture occurs if the grain size is sufficiently fine. This behaviour is discussed on the basis of dislocation theory.

Testing Cold-Drawn Wire for Fracture when Knotted. D. A. Kal'ner. (*Zavodskaya Laboratoriya*, 1956, **22,** (3), 323–325.) [In Russian]. On the basis of experiments with cold-drawn wire of several types of steel the conditions under which tests with knotted specimens give reliable indications of plasticity reserves and tendency to ageing are considered. Results of tests with un-knotted specimens are compared.

The Tensile Strength of Light Steel. E. W. Williams. (*Canad. Min. Met. Bull.*, 1957, **50,** Mar., 147–153). "Con-

trolled density steel" prepared by direct reduction of oxide powders with sintering and carburizing has been examined for mechanical properties. Preparation, metallography and testing are outlined and tensile strength, the effects of quenching, phosphorus, alloying elements and other factors are discussed.

Tensile Fracture of Ductile Materials. H. H. Bleakney. (*Canad. Metals*, 1957, **20**, Feb., 60, 62–64). A discussion is presented of the difference between theoretical and observed cohesive strengths. Al and mild steel were studied and the fracture mechanisms observed are described. The data suggests that no test is available to give true stress at fracture.

Rigidity of Testing Machines and Its Effect on the Falling Part of Extension and Bending Diagrams. F. S. Savitskiĭ and B. A. Vandȳshev. (*Zavodskaya Laboratoriya*, 1956, **22**, (6), 717–721). [In Russian]. The rigidities of seven Soviet and foreign testing machines were determined with the aid of an hydraulic mechanism and the data together with bending and tensile test results, were used to assess the effect of rigidity on the falling part of the stress-strain diagram. The rigidities of all the machines tested were found to be sufficient for correct recording of tensile diagrams even with plastic metals (such as soft steel at room temperature). Special attention was given to the further influence of crack formation in test-pieces. It is recommended that testing machine rigidities should be specified by the makers.—S. K.

Mechanical Tests with Variable Pliability of Loading. T. K. Zilova and Ya. B. Fridman. (*Zavodskaya Laboratoriya*, 1956, **22**, (6), 712–717). [In Russian]. It is shown that in addition to the usual loading conditions it is essential to take into account the pliability of the loading arrangement and the test piece. The effects of pliability on test results under various conditions are examined and the possibility of this affecting the diagram of deformation in tension is indicated.

Influence of Tensile and Compression Stresses on the Magnetic Permeability of Steels. G. Vidal and P. Lanusse. (*Rev. Mét.*, 1957, **54**, Feb., 126–134). An Iliovici permeameter has been adapted to a machine designed to submit the same test piece to tensile or compression stresses. The results indicate that irregularities observed in the magnetic induction—stress curves do not normally exist in the case of compression but under tension they appear. The anomalous behaviour is not related to modifications of state in the steel but is magnetic in origin and can be explained in terms of the magnetostriction curves of the stressed specimen.—B. G. B.

Gaseous-Discharge Indicator for Small Displacements. L. A. Goncharskiĭ and M. Ya. Mulyarov. (*Zavodskaya Laboratoriya*, 1956, **22**, (4), 496–498. [In Russian]. In the sensitive instrument described the deformation of, e.g. a metallurgical test-piece, is transmitted via a flexible seal to one electrode of a gaseous-discharge tube. The voltage required to start the discharge is measured, indicating the electrode gap and hence the displacement.—S. K.

An Investigation of the Fracture of Light Rolled-Steel Sections. K. J. Pascoe. (*Brit. Welding J.*, 1957, **4**, Mar., 133–146). Mild steel and three structural alloys in notched sections containing weld runs were loaded in bending at 0° and − 60° C to study cleavage fracture, and correlated with notch-bend tests. It was found that machined notches, and welds laid with excessive current, would start fractures, and if below the ductile-brittle transition point cleavage occurred with little ductile bending. Repeated loading produced failure below the load producing failure at a single loading. Results are tabulated, and the repeated-loading machine used is described.

Method of Testing Steel for Cold Bending. P. I. Sokolovskiĭ and L. I. Gladshteĭn. (*Zavodskaya Laboratoriya*, 1956, **22**, (3), 331–333). [In Russian]. The relation between test results in the cold-bending of steel round mandrels and tensile tests is discussed, and experiments in which the deformation of specimens was studied are described. The longitudinal distributions of plastic deformation in specimens of O.H. rimming steels bent round mandrels 5, 13, and 26 mm in diameter were determined and in further experiments crack formation was stimulated by preliminary deformation of 2 or 15%. Maximal values of local elongation and concentrated deformation at failure of the test-pieces are tabulated. A graph for choosing mandrel diameters for different test-piece thicknesses is presented.—S. K.

Relation between Static and Impact Work in the Bending of Menager Test-Pieces. (*Zavodskaya Laboratoriya*, 1956, **22**, (3), 365–368). [In Russian]. Some of the main deductions and recommendations made by Yu. D. Guzevich and Yu. P. Sataev [*Idem*, **21**, (6)] as well as the method of reaching them are strongly criticized by E. M. Shevandin.—S. K.

Influence of External Energy on Brittleness in Bend Tests. A. Almar-Naess. (*Brit. Welding J.*, 1957, **4**, Feb., 88–102). It has been shown that internal energy release is the maintaining force in brittle fracture. It is now found that the external energy released has the same qualitative effect as internal energy, that transition temperatures of notched bend specimens can be influenced by the type of testing machine and that the likelihood of brittle fracture is dependent on the stiffness of the entire structure.

Circular Cast Iron Plates. C. Dyrbye. (*Acta Polytechnica*, 1956, **3**, (7), pp. 12). Rupture loads are calculated, and compared with bending tests, from plastic and elastic theories, the measured values falling midway between the two and either can be used with a correction factor.

Torsional Properties of Steels at High Rates of Strain. P. G. Jones and T. J. Dolan . (*Illinois Univ. Eng. Exper. Sta. Bull.*, No. 438, 1957, **54**, pp. 23). Strain rate, temp., type of notch and size of specimen of 4 steels were studied. Torsional strains were obtained from a fly-wheel at angular velocities of 0·5, 27 and 1300 r.p.m. at − 100°, + 700° F and room temperature. Continuous records were obtained. Increase of strain rate caused no appreciable change in strength but did decrease ductility and total energy absorption. Temperature increase decreased strength and energy at low rates and increased ductility at high rates. Size effect increased with rise in speed and fall in temp.

Influence of the Dimensions of Grey-Iron Test-Pieces on Compression Strength. S. S. Nekrasov. (*Zavodskaya Laboratoriya*, 1956, **22**, (4), 484–485). [In Russian]. The compression strength of grey iron was found to depend both on the test-piece diameter (with diameter/length = constant) and on the diameter/length ratio. For comparable results test-pieces of identical dimensions must be used.

Some Cases of Stress Due to Temperature Gradient. D. J. Bergman. (*Trans. Amer. Soc. Mech. Eng.*, 1956, **78**, July, 1011–1019). A qualitative analysis is given of the stress due to a maintained temperature differential in a flat bar in both free and restrained conditions and compared with the cases of a flat plate and a thick pipe. Practical cases are reviewed including a pipe heated from the inside, heater tubes with high internal pressure and high heat inputs, localized heating during welding of corrosion liners, differential expansion due to non-compatible welds, vapour barriers through an insulation wall and the temperature-transition skirt for hot vessels.—D. H.

Rational Application of Sach's Method for the Determination of Residual Stresses in Solid and Hollow Cylinders. L. A. Glikman and A. N. Babaev. (*Zavodskaya Laboratoriya*, 1956, **22**, (4), 466–472). [In Russian]. Some modifications of Sach's method for obtaining a complete residual-stress pattern on a single cylinder (disc) by successive turning and boring, are presented. The complete pattern is obtained by interpolation of stresses in the inner zone. This is especially advantageous when the pattern in the outer zone of the cylinder (disc) or in the outer or inner layers of a hollow cylinder is characterized by a high residual-stress gradient. The modifications developed do not complicate Sach's method. For deformation measurements wire strain-gauges were found to be as accurate as optimeters and more convenient.

Methods of Notch-Toughness Testing of Ship Plate Steel. N. Christensen and B. Augland. (*Brit. Welding J.*, 1957, **4**, Feb., 69–87). Impact tests on 14 ship plate steels and one Mo steel are reported, also static tensile bend and slow bend tests. Energy absorption, lateral deformation and fracture appearance are correlated. An approach to service failure conditions has been made by introduction of an embrittled zone beneath a sharp indentation. The arrest of cracks originating at such notches has been evaluated as a criterion of fracture propagation transition. The significance of temp. range of fracture arrest and its relation to elastic energy is discussed. There is evidence that highly ductile fully killed steels may be unable to arrest brittle fracture in proportion to their ability to resist fracture initiation.

Further Studies on Stainless-Steel Hot Cracking. P. P. Puzak and H. Rischall. (*Welding J.*, 1957, **36**, Feb., 57s–61s). Test results are reported which were obtained with steel type 347 (previously established to be most susceptible to hot

cracking) and type 304 (not prone to hot cracking). The hot ductility and strength developed in high-temperature tensile tests conducted in the temperature range below the repair of normal incipient melting were determined to be of low order only for those heats which exhibited base-metal hot cracking.—U. E.

Scoring Characteristics of Thirty Eight Different Elemental Metals in High Speed Sliding Contact with Steel. A. E. Roach, C. L. Goodzeit and R. P. Hunnicutt. (*Trans. Amer. Soc. Mech. Eng.*, 1956, **78**, Nov., 1659–1667). A study of the 'score resistance' of 38 elemental metals tested as slider bearings against hardened and tempered medium-carbon steel discs of hardness 28–30 R_c is reported. Metals having the best score resistance against steel are the B. subgroup metals which are either insoluble in iron or form intermetallic compounds with iron.—D. H.

Frictional Characteristics and Surface Damage of Thirty-Nine Different Elemental Metals in Sliding Contact with Iron. C. L. Goodzeit, R. P. Hunnicutt, and A. E. Roach. (*Trans. Amer. Soc. Mech. Eng.*, 1956, **78**, Nov., 1669–1676). An investigation of the friction and surface-damage characteristics of iron in sliding contact with various elemental metals without lubrication is described. No simple direct relationship was found between the coefficient of sliding friction and the characteristics of the mutual transfer between metals. The surface damage characteristics are related to the relative hardness of the metals in sliding contact, their mutual solubility and their ability to form intermetallic compounds.

Materials for Small, Oil-Free Bearings. H. S. White. (*Mat. Methods*, 1957, **45**, Apr., 135–139). Coefficients of friction against steel shafts are given.

Hydrogen Permeation through Steel during Abrasion. R. C. Frank and D. E. Swets. (*J. Appl. Phys.*, 1957, **28**, Mar., 380). A brief note on the penetration of hydrogen into metal subjected to wet or dry abrasion where the hydrogen is derived from the water or atmospheric water vapour.

Symposium on Optical and Surface Methods of Non-Destructive Testing. I, S. Tolansky; **II,** R. W. B. Stephens; **III,** O. S. Heavens. (*Brit. J. Appl. Phys.*, 1957, Supp. 6, S5–S12). An account of surface examination is given including study of etch pits and hardness indentations by interferometric methods, optical shadow casting, polarization techniques, reflectivity, refractivity, and surface roughness studies.

Optical Fluorescence in Non-Destructive Testing. W. A. Runciman. (*Brit. J. Appl. Phys.*, 1957, Supp. 6, S34–S36). A review.

Symposium on Principles of Penetrant Methods. I, H. L. Carson; **II,** W. G. Cook; **III,** R. Schnurmann (and Discussion). (*Brit. J. Appl. Phys.*, 1957, Supp. 6, S18–S23). Crack detection, using dyes or fluorescent inks, is reviewed.

Magnetic Properties and the Structure of Metals. W. Sucksmith. (*Brit. J. Appl. Phys.*, 1957, Supp. 6, S24–S27). A review.

Reproducing the Properties of Alnico Permanent Magnet Alloys with Elongated Single-Domain Cobalt-Iron Particles. F. E. Luborsky, L. I. Mendelsohn, and T. O. Paine. (*J. Appl. Phys.*, 1957, **28**, Mar., 344–351).

An Investigation of the Structural Conditions in Steel Bearing-Balls. I. Berz, K. F. Hale, and C. Wainwright. (*Metallurgia*, 1957, **55**, Mar., 121–124). Methods are described which are capable of application as non-destructive physical tests of steel bearing balls. These tests are (*a*) a strength test by spinning in a rapidly rotating magnetic field, (*b*) a high frequency electrical test for surface conditions, (*c*) X-ray diffraction examination. The X-ray survey showed the existence of polar areas which could be related to residual flow-line markings seen on etched sections. A correlation between the X-ray and the high frequency electrical methods was found.—B. G. B.

An Investigation of the Eddy-Current Anomaly in a Low-Silicon Sheet Steel. H. Aspden. (*Proc. Inst. Elect. Eng.*, 1957, **104C**, Mar., 2–7).

The Determination of Current-density Voltage-curves with an Electronic Potentiostat with Reference to Electrolytic Insulation. W. Koch and H. Lüdering. (*Arch. Eisenhüttenwesen*, 1957, **28**, Apr., 201–206). In the instrument described, the electrode potential is pre-determined and is maintained substantially independent of voltage in order to avoid the jumps in potential that would occur in steels capable of being rendered passive, and to enable a smooth transition from cathode loading to anode loading of the electrode. Current-density voltage-curves in citrate electrolyte have been recorded for various steels with this instrument.—L. J. L.

Electronics. C. A. Vossberg. (*Iron Steel Eng.*, 1957, **34**, March, 93–97). The article describes a number of practical applications of electronics to control problems in the iron and steel industry.—M. D. J. B.

The Interpretation of Ultrasonic Testing for Finding Faults in Steel. E. Marianeschi and T. Tili. (*Met. Ital.*, 1957, **49**, Apr., 275–289). [In Italian]. The authors describe the basic theoretical aspects of ultrasonic testing. The effects of acoustic impedence, reflection, scatter, diffraction, the influence of the obstacle size etc. are discussed in detail. The results of experimental tests are given and explained—the agreement between theory and practice is shown. Finally the authors propose a new method based on the use of different ultrasonic frequencies for distinguishing between internal cracks and slag inclusion. (14 references).—M. D. J. B.

The Interpretation of Ultrasonic Tests on Steel Products. A. Michalski and H. Krächter. (*Arch. Eisenhüttenwesen*, 1957, **28**, Apr., 213–220). With the ultrasonic reflection method, a standard scale can be developed for the records produced by the instrument from the ratio of the height of the intermediate echo to the height of the back wall echo, the characteristics of the luminous screen image, and the permeability of the ultrasonic waves. Factors influencing echo height are explained; examples of the application of the scale to rods, forgings, and metal sheet are given, and the limits of applicability of the method are discussed.—L. J. L.

Automatic Flaw Detection by Ultrasonics. (*Automation Prog.*, 1957, **2**, June, 276–277, 297). A brief account of the Autosonics system.

Ultrasonic Automation—The Autosonic System. (*Instrument Practice*, 1957, **11**, Apr., 353–356). The system developed by Kelvin and Hughes (Industrial) Ltd. is described with brief specifications and an account of uses.

Autosonics: New Kelvin-Hughes Automatic Testing Plant. (*Iron Steel*, 1957, **30**, Apr., 130, 149). Details are given of the recently developed Kelvin-Hughes automatic ultrasonic inspection ("Autosonics") plant, which can operate without supervision and automatically record and mark defects.—G. F.

Lamination Detection. New High-Speed Inspection Equipment. (*Brit. Steelmaker*, 1957, **23**, May, 142–143). An account of the Kelvin and Hughes (Industrial) Ltd. "Autosonics" equipment is given.

Ultrasonic Testing of Heavy Steel Products. I. M. Mackenzie and R. Kennedy. (*Engineering*, 1957, **183**, May 24, 652–655). A review of techniques and applications, particularly to the detection of pipe in large ingots, examination of forgings, bars and blooms, detection of hairline cracks and laminations. Echo transmission and multiple-echo techniques are discussed.

X-Ray Attenuation Coefficients From 10 kev to 100 Mev. G. W. Grodstein. (*Nat. Bur. Standards Circular* 583, 1957, Apr. 30, pp. 54). Iron and other metals, elements and compounds are tabulated with a discussion of processes and phenomena.

The Potentialities of X-Ray Diffraction Studies in Non-destructive Testing. J. R. Rait and H. S. Peiser. (*Brit. J. Appl. Phys.*, 1957, Supp. 6, S13–S17). A review with an account of conventional methods and their limitations is given, and high-voltage diffraction studies of steels are given showing effects of fatigue.

Use of Gamma Rays for Continuous Inspection. Detection of Cavities in Semi-finished Products at Rolling Temperature. A. Kohn. (*Iron Coal Trades Rev.*, 1957, **174**, May 3, 1015–1018; May 10, 1083–1086; May 17, 1147–1151). Theory, characteristics of γ-rays from ^{60}Co, collimation, the absorption coefficient of iron, and factors in sensitivity and a description of an installation in a French rolling mill and the test conditions and results achieved are reviewed.

Nuclear Metallurgy (Vol. III). (*Amer. Inst. Min. Met. Eng. Special Report No.* 3, 1956, pp. 54). A report of a Symposium on the effects of radiation on metals at Cleveland, Ohio. **Theory and Mechanism of Radiation Effects in Metals.** G. H. Vineyard. (1–12). (49 references). **Irradiation Effects on Physical Metallurgical Processes.** D. E. Thomas. (13–30). Cu–Fe and austenitic stainless steel are discussed as well as non-ferrous systems. Phase transformations and order–disorder processes in non-ferrous systems are reviewed. **Irradiation Effects in Reactor Materials.** D. S. Billington. (31–54). Steels are included.

Use of Ultrasonics in Non-Destructive Testing. S. A. Lund.

(*Dansk Tekn. Tidsskr.*, 1957, **81,** Feb., 43–49). An account of available generators and their uses is given. Testing of welds and flaw detection in large components is described and the kinds of traces obtained are shown.

The Present State of Non-Destructive Testing by Means of X-Rays in Czechoslovakia and Abroad. M. Baimler. (*Sborník Defektoskopie*, 1956, 9–50). [In Czech]. The performance and design of industrial X-ray generators working with accelerating potentials in the range of 50–2000 kV, as well as betatrons operating with potentials up to 31 MeV are reviewed and discussed. Numerous photographs of equipment are shown.

Theoretical Evaluation of the Scope of Non-Destructive Testing by Means of X-Rays. F. Kohl. (*Sborník Defektoskopie*, 1956, 3–7). [In Czech]. A survey is made of the basic principles of X-ray defectoscopy and of the physical aspects of generating and detecting the radiation.—P. F.

Where New X-Ray Techniques Solve Old Shop Problems. (*Iron Age.*, 1957, **179,** Feb. 14, 119–121). The X-ray diffractometer is described, it is a spectrometer with radial detector which produces an inked record of location and intensity. The kind of question it can answer is listed.

Detectors of Radiations from Radioactive Isotopes and their Use in Thickness Gauges in the Metal Industry. M. Krejčík. (*Sborník Defektoskopie*, 1956, 119–133). [In Czech]. Detectors and counters are discussed, and basic theoretical principles are outlined. An accuracy of about 3% has been obtained by the author in measuring steel strip width in the works in the process of production.—P. F.

Methods of Measuring Ionising Radiations. J. Oppelt. (*Sborník Defektoskopie*, 1956, 133–136). [In Czech].—P. F.

The Use of Radio-Isotopes for Gamma-Ray Defectoscopy. G. Kuba. (*Sborník Defectoskopie*, 1956, 85–111). [In Czech]. Gamma-ray defectoscopy is surveyed, and problems involved in handling isotopes, as well as the design of appropriate containers and carriers, are considered.—P. F.

Nuclear Metallurgy (Vol. III). G. H. Vineyard, D. E. Thomas and D. S. Billington. (*Amer. Inst. Min. Met. Petroleum Eng., Special Report Series* **No. 3.** 1956, pp. 54). **Theory and Mechanism of Radiation Effects in Metals.** (1–12). Processes are reviewed with values for particle ranges, transition energies from ionization stage to elastic collision stage (in graphite, Al, Cu, Au), mean energy of primary knock-ons. Determination of threshold energy and number of displaced atoms is described, and the effect of defects considered. An appendix on the calculation of cascade processes is added. (49 references). **Irradiation Effects on Physical Metallurgical Processes.** (13–30). Diffusion, precipitation from solid solution in systems including copper–iron and austenitic stainless steel, phase transformations and order–disorder processes (in non-ferrous systems) are discussed. (49 references). **Irradiation Effects in Reactor Materials.** (31–54). Selection of materials is considered according to function in the reactor and problems of experimentation in the field. Effects in U and its alloys, in graphite and other moderators, in shielding material and control rods and in structural materials are then reviewed. Changes in steel, Armco iron and other metals are tabulated and particularly effects in ASTM A-212 Grade B C–Si steel. (77 references).

Internal Friction of Turbine-Blade Cr-steels. J. Vodsedálek. (*Materiálový Sborník*, 1956, 5–25). [In Czech]. Damping capacities of a number of Czechoslovak and foreign steels were studied in the range 0–800° C under conditions of cyclic stressing. The influence of chemical composition, structure, heat-treatment and magnetic properties on the damping are analysed. The criterion of quality used is the product (fatigue limit) × (damping at the fatigue limit). A number of steels are classified on that basis. Effects due to pre-stressing are considered.—P. F.

Properties of Metals at Elevated Temperatures. N. J. Grant, H. C. Chang, F. C. Monkman and P. E. Price. (*PB* 118662, 1954, pp. 4; *TIDU List* 703). A quarterly progress report on intercrystalline cracking, a vacuum crucible furnace for study of the Fe–Cr–Ni system, and the effect of microstructure on the high-temperature stress-rupture strength of stainless steels with particular reference to trace elements.

Deformation Resistance of High Temperature Steels at 800, 900 and 1000° C. S Terai. (*Sumitomo Metals*, 1956, **8,** Oct., 21–28). [In Japanese]. Low-carbon, 21/4 Cr/Mo, 17 Cr, 25 Cr, 18 Cr 8 Ni, 16 Cr 13 Ni 3 Mo and 22 Cr 12 Ni 1·5 Si steels (analyses given) were tested at 10–0·1%/h at the three temperatures and the results plotted on log/log paper. $\alpha \rightarrow \gamma$ transformation seemed to occur and γ is stronger than α. Comparison with creep data is made and with high-temperature plastic working.

Properties of Steel at Elevated Temperatures. (A review of the Literature Published in 1955). A. Krisch. (*Stahl u. Eisen*, 1957, **77,** Jan. 24, 108–112). The author continues his review and covers the effect of composition on creep properties.

Properties of Steel at Elevated Temperatures. (A Review of the Literature Published in 1955). A. Krisch (*Stahl u. Eisen*, 1957, **77,** Feb. 7, 175–178). The author concludes his review with: Effect of structure on creep properties of steel; a short note on grey cast iron; and non-ferrous metals including powder metallurgy, mainly cemented carbides.—T. G.

Further Studies on Stainless-Steel Hot Cracking. P. P. Puzak and H. Rischall. (*NRL Report* 4861, 1956, pp. 10). Types 304 and 347 steels appear to crack owing to grain-boundary liquation of segregates. Further studies were made by thermal analysis and hot-ductility was measured just below the incipient melting region. This was low for metals exhibiting hot-cracking, and grain-boundary liquation areas were observed. Analysis and X-ray studies suggest a eutectic-type mixture of CbC and CbN with minor amounts of Mn, Si and Fe is the probable segregate involved.

Diffusion in Metals. T. J. Tiedema and A. Bolk. (*Metalen*, 1957, **12,** Feb. 28, 60–64). [In Dutch]. This paper deals with some recent developments in the field of interpreting concentration-penetration curves and the authors go on to describe briefly the Kirkendall effect and Darken's method of calculating partial diffusion coefficients. These problems are elucidated on the basis of the Cu–Ni system and to conclude a survey is given of the aspects of grain boundary diffusion and the influence on this phenomenon of the boundary structure. (8 references).—F. R. H.

Contribution to the Study of Diffusion of Carbon in Iron. P. Brissonneau. (*Compt. Rend.*, 1957, **244,** Mar. 10, 1341–1344). A magnetic study is reported suggesting that lattice imperfections are directly responsible for the dispersion found.

Diffusion and Solubility of Iron in Germanium. A. A. Bugai, V. G. Kosenko and E. G. Miselyuk. (*Zhur. Tekn. Fiz.*, 1957, **27,** Jan., 210–211. The range 750–940° C is covered.

Surface Tension of Iron, Cobalt and Nickel at 1550° C. P. Kozakevitch and G. Urbain. (*Compt. Rend.*, 1957, **244,** Jan. 14, 335–337). The hanging drop method was used. The value for Fe was 1835 dyne/cm^{-1}.

Hydrogen Content and Lattice Stresses as Causes of Flaking. K. Kreitz. (*Stahl u. Eisen*, 1957, **77,** Feb. 7, 168–169). The author reviews published theories on flaking of steel as caused by hydrogen. He points out that the introduction of the concept of moving dislocations could bridge the gap existing between the hydrogen pressure theory and the transformation-stress theory.—T. G.

Hydrogen Embrittlement of Cold-worked Metals. R. L. Mills and F. J. Edeskuty. (*Chem. Eng. Prog.*, 1956, **52,** Nov., 477–480). A test method in which strips are inserted in a high-pressure cylinder and exposed to hydrogen at 40,000 p.s.i. is described. Steels AISI 303, 316, 347 (from 2 sources), C 1020 and C 1022 were used, compositions and mechanical properties are tabulated. A bend machine is described on which subsequent tests were made and embrittlement assessed.

The Hydrogen Embrittlement of Iron and Steel. W. Rostoker. (*Amer. Iron Steel Inst. Reg. Tech. Meetings*, 1956, 121–134). The effects on ductility, flaking and static fatigue are considered and mechanisms are outlined.

An Investigation of the Effects of Hydrogen on the Brittle Failure of High-Strength Steels. E. R. Slaughter, E. E. Fletcher, A. R. Elsea and G. K. Manning. (*PB* 121357, 1956, pp. 58; *TIDU List* 723). Failures can occur at below 10% of the normal ultimate tensile strength, and brittle fracture is unaffected by variations in composition or structure but decreases as tensile strength of the steel increases. The effect of hydrogen on the martensite reaction is discussed. Methods of testing and results are described.

Delayed Failure and Hydrogen Embrittlement in Steel. R. P. Frohmberg, W. J. Barnett and A. R. Troiano. (*PB* 121085, 1954, pp. 61; *TIDU List* 722). Mechanisms are reviewed and a study of SAE–AISI 4340 steel is reported in which hydrogen was introduced by electrolytic charging. Effects on properties, especially after ageing, are given. A note on bainitic structure is included.

Crack Initiation in Hydrogenated Steel. H. H. Johnson and

A. R. Troiano. (*Nature*, 1957, **179**, Apr. 13, 777). A review of embrittlement phenomena, flaking and crack formation is given in a letter. Electrical resistance measurements are reported and an incubation period was observed. This is compatible with either of the major theories of the phenomena. The kind of test needed is briefly discussed.

Destruction of Steel by Hydrogen. V. Jareš. (*Problems and Perspectives of Czechoslovak Metallurgy and Foundry*, 1956, 48–60). Hydrogen diffuses into steel under higher pressures, and decomposes carbides with formation of methane, producing embrittlement. Failures in service are described and choice of steels for various purposes is indicated. Only partial success has been attained in the regeneration of embrittled steels.

Statistical Analysis of Notch-Toughness Values Obtained on Different Pendulum Testing Machines. A. Schepers and F. R. Licht. (*Stahl u. Eisen*, 1957, **77**, Feb. 21, 218–221). Steels C15 and C60 were tested in a 15- and a 30-kg.m. pendulum testing machine. Both hammers had the same speed at impact. The 15-kg.m. machine gave higher values for higher impact strength than the 30-kg.m. machine, e.g. for steel C15 15·77 as against 15·05 kg.m./cm². For lower impact strength, shown on steel C60, the values were almost identical. The reasons for the discrepancies are analysed and explained.—T. G.

International Research on the Mode of Yield of Weldable Steels C and D. (Study of Shock on Resilience Samples with a V notch). R. Borione. (*Centre Doc. Sidér. Circ. Inform. Tech.*, 1957, **14**, 201–205). Tests carried out by 3 different laboratories on two different steels are reported. The reproducibility and degree of scatter obtained was the principal factor studied.—B. G. B.

Consideration of the Deformation and Rupture of Mild Steel in Relation to the Problem of Sudden Fracture. R. Borione. (*Bull. Cercle Etudes Mét.*, 1956, **7**, Sept., 47–82). Factors affecting the mode of fracture of mild steel are reviewed in the light of published data. Rheotropic embrittlement and brittle fracture are among the factors considered.

Determination of the Work to Fracture in Impact Tests of Parts on a Machine with Vertical Fall of Bob. G. V. Zarochentsev. (*Zavodskaya Laboratoriya*, 1956, **22**, (4), 478–481). [In Russian]. In the method described the fall of an impact testing bob is followed photo-electrically. A grid is attached to the bob so that during the fall its bars pass between a lamp and a photoelectric cell, the resulting impulses being recorded oscillographically. Details and circuit diagrams are given. The photoelectric arrangement enables the energy producing fracture, found from the difference of the squares of the limit velocities to be obtained more reliably than with mechanical linkage. Some possible modifications of the optical system are considered.—S. K.

Multiple Scabbing in Materials. S. Kumar and D. Davids. (*J. Franklin Inst.*, 1957, **263**, Apr., 295–302). Fracture by suddenly applied pressures of high intensity is reviewed, especially by explosions.

Influence of the Rate and Amplitude of Plastic Strains on the Behaviour of Steel Containing Hydrogen. P. Bastien and P. Azou. (*Rev. Mét.*, 1957, **54**, Feb., 93–100). The influence of rate and amplitude of pre-existing plastic deformation on the behaviour of extra-mild steels containing hydrogen has been studied by direct etching and by electrolysis.

Use of Ultrasonic Methods for Studying the Development of Fatigue Cracks in Shafts under Pressed-On Parts. I. N. Ermolov, M. M. Kobrin, and M. F. Krakovyak. (*Zavodskaya Laboratoriya*, 1956, **22**, (6), 724–728). [In Russian]. An account is given of investigations which have shown that ultrasonic detection of the appearance and development of fatigue cracks in shafts under pressed-on sleeves is feasible. Preliminary experiments were carried out using shafts with artificial cracks to determine optimal experimental conditions. The minimal crack-depth detectable (0·5 mm) was found using sleeved shafts subjected to variable-sign bending, the indications of the ultrasonic method being checked by the magnetic-powder method after removing the sleeve. Fatigue curves are presented for sleeved shafts bent in one plane without rotation.—S. K.

Crystallographic Behaviour of a Weakly Carburized Steel During Fatigue Testing at High Temperature (< 800°). R. Jacquesson and J. de Fouquet. (*Compt. Rend.*, 1957, **244**, Apr. 29, 2306–2308). Ugiperval iron samples annealed 1 hr/950° *in vacuo* and electropolished were tested in a furnace in an atmosphere of A. The course of polygonization is reported under various conditions.

Operating Experience of Resonance Type Fatigue Machines. S. I. Yatskevich and N. A. Balabanov. (*Zavodskaya Laboratoriya*, 1956, **22**, (7), 858–860). [In Russian]. An account is given of several years' experience in fatigue testing with resonance-type machines. New methods of securing the machine to the foundations and of measuring deformations are described.—S. K.

Methods of Analyzing Fatigue Fractures. A. M. Zaitsev. (*Zavodskaya Laboratoriya*, 1956, **22**, (4), 472–478). [In Russian]. The development of fatigue fracture occupies a relatively large proportion of the total time to fracture, and the corresponding changes in conditions leave characteristic traces on the fracture material. In this article these traces and their interpretation are considered in some detail. Among effects considered are the quantity of steps or ridges in the fractures and the nature of their distribution, these being governed by the value of the applied stress and the nature of the surface at which they originate. The inferences to be drawn from the distribution of fatigue lines and bright surfaces in the fracture are considered for a fractured crankshaft, and characteristic effects are illustrated.—S. K.

Methods of Statistical Analyses of Results of Fatigue Tests Carried Out on Aeronautical Materials. F. Gatto. (*Métaux-Corrosion-Indust.* 1957, **32**, Jan. 18–37). A comprehensive account of statistical techniques of use in assessing the results obtained in the fatigue testing of metals is presented. A number of examples is given.—B. G. B.

Method of Producing Fatigue Failures in Railway Rails. R. S. Nikolaev. (*Zavodskaya Laboratoriya*, 1956, **22**, (6), 721–724). [In Russian]. A new testing machine for inducing fatigue failures in rails is described and some results obtained are presented and discussed. The rail is subjected to a rolling load, and a special device has been developed for turning the test-piece. Longitudinal horizontal, induced and vertical laminations as well as transverse fatigue cracks in the head are produced by the machine. The effects of inclusions on the behaviour of test-pieces are discussed, those of alumina being especially important since their great hardness prevents plastic deformation; inclusion-filled gas pores are also associated with failures.—S. K.

Simple Machines for the Determination of the Creep Characteristics of Steels and Alloys. A. Gueussier and R. Castro. (*Rev. Mét.*, 1957, **54**, Feb. 101–106). A simple, strongly built creep testing machine designed by Acieries Electriques d'Ugine is described. The test piece is loaded by means of a weight through an amplifying eccentric-lever system placed under the furnace, which is horizontal. A simple temperature control is also described. The whole equipment takes up little floor space.—B. G. B.

Effect of Ceramic Coatings on Creep of Alloys. J. R. Cuthill, J. C. Richmond, and N. J. Tighe. (*Indust. Heating*, 1957, **24**, Mar., 582, 584, 586, 588). Work on a Ni–Cr aircraft alloy is described.

How to Use Creep Data in Design. W. F. Simmons. (*Mat. Methods*, 1956, **44**, Nov., 120–123). Creep in stationary equipment, aircraft components and airframe structures, mainly non-ferrous is discussed.

Apparatus for Measuring the Hardness of Internal Surfaces. I. A. Begagoen and G. K. Tatur. (*Zavodskaya Laboratoriya*, 1956, **22**, (4), 490–492). [In Russian]. The modification of a device for measuring the hardness of internal cylindrical surfaces with a fixed diameter over a wide range of diameters is described. The instrument uses an indentation method, a special microscope being used for its observation.—S. K.

Contribution to the Study of the Problem of Brittle Fractures. W. Soete. (*Rev. Mét.*, 1957, **54**, Jan., 71–79). The efficacy of a single Robertson test is shown and checked in the cases of biaxial external loading and high residual stresses. The influence of metallurgical and constructional factors on the crack arrest temperature has been studied and the results obtained are discussed in detail.—B. G. B.

Cause and Prevention of Brittleness in Steel. (*Mech. World*, 1956, **136**, July 328–329). The cause and prevention of brittleness in steel is discussed in general terms. The testing and inspection of material for susceptibility to brittleness is briefly described.—D. H.

Temper Brittleness of Boron-Treated Steels. S. J. Rosenberg. (*J. Res. Nat. Bur. Stand.*, 1957, **58**, Apr. 175–187). The investigation was carried out to ascertain whether

Ti or Zn had any adverse effect on impact properties of B steels. All samples were hardened at 1600° F/30 min and quenched and tempered at 1200° F/1 hr. The rate of cooling was varied to develop temper-brittleness and an isothermal treatment was also used. Ti, but not Zr (in the small amounts usually present) impaired impact properties.

The Brittle Fracture Strength of Welded Steel Plates. A. A. Wells. (*Trans. Inst. Naval Architects*, 1956, **98**, 296–326). Tensile tests on 1 in. mild steel plates with notched butt welds parallel to the direction of pull at ± 15° C are reported. Brittle fracture occurred below 4° C. An explanation of the results is offered.

Thermal Cycling Test of a Hot Spot on a Vessel. P. N. Randall and H. A. Lang. (*Trans. Amer. Soc. Mech. Eng.*, 1956, **78**, July, 1003–1010). The results of laboratory tests simulating the effect of repeated heating and cooling cycles on a hot spot on a catalytic-cracker regenerator shell are given. After 50 cycles of heating to 900° F at the spot centre for 1–2 days and cooling to near room temperature, radiographic and metallographic examinations of the specimen were carried out and mechanical properties determined. No evidence of cracking or deterioration of the structure was observed.—D. H.

Fatigue Failure—A Structure Study. A. A. Krishnan and K. D. Maji. (*J. Sci. Ind. Res.*, 1957, **16B**, 105–116). Studies on copper show the development of cracks and mechanisms are discussed.

Fatigue of Metals Under Combinations of Stresses. W. N. Findley. (*Trans. Amer. Soc. Mech. Eng. preprint*, 56–A-74, pp. 11). A machine for combined bending and torsion fatigue testing is described. Repeated shear stress seems to be the cause of fatigue fracture and other modifying influences are mentioned. Theories of fatigue require correction for these influences. A better understanding of the origin of fatigue at the atomic level is required.

Effect of Plastic Fatigue on Pressure Vessel Materials and Design. L. F. Kooistra. (*Welding J.*, 1957, **36**, March, 120s–130s). Data have been correlated to establish the parameters governing pressure-vessel design with respect to plastic fatigue characteristics of the material. Testing methods have been evaluated for determining " Plastic endurance " as a characteristic of the material itself, as well as methods for establishing the " effective strain concentration factor " of various plate surface conditions, notches, attachments, and the effect of welding.—U. E.

Fatigue Strength of Specimens Containing Preformed Fatigue Cracks. N. E. Frost. (*Engineer*, 1957, **203**, June 7, 864–867). Small cracks were formed in round bar specimens by rotation bending and direct fatigue tests were then carried out. A 0·15% C mild steel, EN26, and an Al alloy were used.

Fatigue Strength of Specimens Containing Cracks. N. E. Frost and C. E. Phillips. (*Proc. Inst. Mech. Eng.*, 1956, **170**, (21), 713–725; discussion, 726–745). Tests on precracked specimens of mild and Ni–Cr steels and non-ferrous alloys are described. Explanations are suggested for crack strength, and crack propagation is discussed. The strength-reduction factor of a crack is independent of its size, specimen diameter, formation conditions, and type of loading when formed in the manner used in this study.

Fatigue Strength of Metals under Alternating Stresses of Varying Amplitude. T. Nishihara and T. Yamada. (*Mem. Fac. Eng. Kyoto Univ.*, 1956, **18**, July, 172–208). Criteria for fatigue damage and formulæ predicting fatigue limits are established. Three types of tests were used on notched and unnotched low and high carbon steels and duralumin. It is concluded that the analysis given correctly predicts fatigue lives and fatigue limits.

Delayed-Yield Time Effects in Mild Steel Under Oscillatory Axial Loads. R. O. Belsheim. (*Trans. Amer. Soc. Mech. Eng., preprint*, 56–A-108, pp. 7). Apparatus for loading specimens with superimposed static and vibratory loads is described. Results were in agreement with other published values. (23 references).

Steady-State Creep through Dislocation Climb. J. Heertman. (*J. Appl. Phys.*, 1957, **28**, Mar., 362–364). An equation is obtained with power law stress dependence at low stresses.

On the Activation Energy of High Temperature Creep in Metals. P. Feltham. (*Phil. Mag.*, 1957, **2**, May, 584–588). Study of the activation volume or " stress-concentration factor " in the second-stage equation for creep is reported and interpreted in terms of dislocation theory.

Creep Properties from Short Time Tests. E. E. Underwood. (*Mat. Methods*, 1957, **45**, Apr., 127–129). A straight line relationship between creep rupture stress and hardness is determined. Time or temperature can be determined when either is fixed.

Creep Characteristics of Type 347 Stainless Steel at 1050° and 1150° F in Tension and Compression. M. J. Manjoine. (*Trans. Amer. Soc. Mech. Eng., preprint*, 56–A-40, pp. 8). Heavy wall tubing was studied in plastic flow. Precipitation induced by strain and temperature causes an extended period of high resistance, longer for compression than for tension.

Creep Damage in a Cr–Mo–V Steel as Measured by Retained Stress Rupture Properties. M. H. Jones, D. P. Newman, and W. F. Brown jun. (*Trans. Amer. Soc. Mech. Eng.*, 1957, **79**, Jan., 117–126). Progressive deterioration was measured at 1000° F and the damage was shown to be both strain and time-sensitive. Mechanism is suggested and it is shown that reheating is able to repair any damage resulting and that creep damage is associated with a time–temperature–dependent structural change in the alloy which is also associated with notch rupture sensitivity.

Validity of Time-Compensated Temperature Parameters for Correlating Creep and Creep-Rupture Data. F. Garofalo, G. V. Smith, and B. W. Royle. (*Trans. Amer. Soc. Mech. Eng.*, 1956, **78**, Oct., 1423–1434). The time-compensated temperature parameters suggested by Larson and Miller, Manson and Haferd and Dorn et al. are discussed with relation to time of creep rupture and to minimum and average creep rates. The ' master ' curve is discussed in terms of experimentally determined relationships. In checking the validity of these parameters using experimental results for several ferritic and austenitic steels, it is found that these parameters are not single-valued functions of the initial stress over wide ranges in stress and the existence of a ' master ' curve is doubtful.

Railway Wheel Testing. B. B. Hundy and J. C. Fleming-Smith. (*Rail Steel Topics*, 1956, **3**, Autumn, 25–40). An illustrated account of the methods used and defects encountered is given.

Report of the Investigation of Two Generator Rotor Fractures. C. Schabtach, E. L. Fogleman, A. W. Rankin. and D. H. Winne. (*Trans. Amer. Soc. Mech. Eng.*, 1956, **78**, Oct., 1567–1584). The paper describes the various phases of an investigation into the causes of failure of generator rotors, and includes a study of design features and manufacture and characteristics of rotor materials.—D. H.

Investigation of Large Steam-Turbine Spindle Failure. H. D. Emmert. (*Trans. Amer. Soc. Mech. Eng.*, 1956, **78**, Oct., 1547–1565). A full report is given of the investigation into the cause of failure of a low-pressure spindle of a cross-compound steam turbine which burst during an overspeed-trip test. It was concluded that the initiating cause of the failure was flakes or thermal cracks developed in the shaft during heat treatment.—D. H.

Report of the Investigation of the Turbine Wheel Fracture at Tanners Creek. A. W. Rankin and B. R. Seguin. (*Trans. Amer. Soc. Mech. Eng.*, 1956, **78**, Oct., 1527–1546). The authors discuss the steam turbine wheel fracture which occurred at Tanners Creek early in 1953. The major parts of the investigation are presented together with the general conclusions.—D. H.

Automatic Steel Inspection. (*Optima*, 1957, **5**, Mar., 42–43). A device for testing the hardness of steel drill stems by magnetic means is described.

Further Developments in the Precision Measurement of Diamond Indentors for Rockwell Hardness Testing. F. R. Tolmon and J. F. Hall. (*Microtecnic*, 1957, **11**, (1), 36–46). NPL work is described.

The Evaluation of Non-circular Brinell Impressions. W. Wepner. (*Arch. Eisenhüttenwesen*, 1957, **28**, 237–239). According to DIN 50 351 non-circular, and in particular elliptical, Brinell hardness impressions are evaluated by determining the diameter in two directions at right angles and calculating the hardness from the mean value. The possibility of errors is examined.—L. J. L.

On the Jominy Test. T. Furukawa. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Mar., 205–208). [In Japanese]. The influence of surface grinding on Jominy hardenability was investigated for several steels. All grinding conditions were kept constant except for the depth of cut, which was varied between 10 and 50 μ. The influence of this was very marked, particularly at the quenching end; it is explained in terms of

changes in the surface layer caused by the heat of cutting and distortion. (14 references).—K. E. J.

Hardenability Test for Deep Hardening Steels. C. M. Carman, D. F. Armiento, and H. Markus. (*Metal Progress*, 1957, **71**, May, 77–80). A test-piece is austenitized uniformly in a tube furnace and then lowered and the protruding end cooled with a jet of water while the furnace is being cooled at a rate corresponding to the position of the test piece. Rockwell values are then measured.

Temper Brittleness. (*Mech. World*, 1957, **137**, Mar., 135–136). A review of present conceptions with a distinction from "blue brittleness." Methods of prevention are considered.

Brittle Fracture of Sheet Piling. W. A. Morgan and R. C. A. Thurston. (*Metal Progress*, 1957, **71**, Mar., 86–91). Several failures of sheet piling in Canadian waters have been noted. Sheet piling is normally rolled from structural steel and the transition temperature, where fatigue cracks or corrosion pits lead to brittle fracture, is above the water temperature. The results of tests carried out on steel sheet piles are given and it is considered that a better grade of steel should be specified for the piles than hitherto.—B. G. B.

The Nominal Cleavage Strength of Steel and its Importance for Welded Structures. S. M. Norén. (*Shipbuilder and Marine Engine Builder*, 1957, **64**, Mid-Apr., 244–247). Brittle fracture, testing methods, NC curves and their interpretation are discussed.

Cracking Tendency in Heavy High-Temperature Steam Piping. H. M. Soldan and C. R. Mayne. (*Metal Progress*, 1957, **71**, Mar., 78–85). The development of cracks in steam piping is reviewed in detail. A stabilized 18–8 stainless steel (type 347) is the conventional material for high temperature steam service in America. Occasional cracks alongside welds may be avoided by using metal and welding electrodes which have a 60% reduction in area or greater when cooled from 2400° F and tested at 2300° F. 16–8–2, Cr–Ni–Mo alloy is satisfactory for this purpose.—B. G. B.

Hydrogen Embrittlement of Steel and its Relation to Weld Metal Cracking. H. G. Vaughan and M. E. de Morton. (*Brit. Welding J.*, 1957, **4**, Jan., 40–61). Theories of embrittlement are discussed and initiation of fractures in weld metal has been related to effects of H_2 in steel. Microscopy shows the pressure build up at matrix interfaces and cleavage induced by the resulting strains. Failures in tensile specimens are predominantly of cleavage type. It is found that in presence of cathodically introduced H_2 the upper and lower yield-point phenomenon disappears and is replaced by a single yield-point. The work hardening curve steepens, the strain-ageing tendency due to N_2 is reduced, but if time for diffusion is insufficient, H_2 produces a similar tendency. If H_2 is allowed to diffuse out normal properties return. (35 references).

The Normal Adhesion of Clean Metals and Its Relation to Friction. G. W. Rowe. (*Abstracts of Dissertations, University of Cambridge*, 1953–1954, 249–250).

Extreme Temperature Tests Show How Metals Wear. (*Iron Age*, 1956, **178**, Dec. 6, 135). The problem of wear caused by friction between two metals in the temperature range − 100° to 2000° F has been studied by Teleflex Inc., North Wales, Pa. An outline is given of some of the tests from which a phosphor-bronze flexible cable, with self-lubricating properties and capable of withstanding temperatures up to 600° F, was developed.—G. F.

Arrangement for Wear Testing according to the "Shaft-Fractional Bush" Scheme. E. S. Nevezhin. (*Zavodskaya Laboratoriya*, 1956, **22**, (3), 364). [In Russian]. A very simple and reliable variant of the wear-testing method in which a shaft of one of the materials of the pair rotates in contact with a section of a bush of the other is described. The bush is built up of sections of rings taken from a suitable range of diameters, and held in a special frame.—S. K.

Contribution to the Study of Magnetic Sorting Equipment. R. Loude and F. Maratray. (*Centre Doc. Sidér. Circ. Inform. Tech.*, 1957, **14**, (2), 367–414). Factors influencing the magnetic properties of ferromagnetic materials are first reviewed. A comprehensive study of the magnetic properties of steels is presented. Commercial equipment used for testing and sorting steels are considered. Copies of oscilloscope traces indicating the variations in B-H curves for different steels, and different test conditions are reproduced in considerable numbers.—B. G. B.

Powerful Magnetic Flaw-Detectors with Continuous Regulation of the Current. P. G. Mikhnevich. (*Zavodskaya Laboratoriya*, 1956, **22**, (7), 860–862). [In Russian]. Two flaw-detectors developed in recent years in which large (up to 10,000 amp.) currents can be obtained are described. The Instruments are the "UMD 9000" and the "MSA 9000".

Coercive-Force Meter with a Magnetic Shunt and Drive. I. P. Vyazovoĭ. (*Zavodskaya Laboratoriya*, 1956, **22**, (3), 344–345). [In Russian]. A circuit diagram and brief description are given of an improved design of coercive-force meter with higher reliability and stability of readings. The arrangement provides for strict constancy of the duration of magnetization and of the interval between the magnetizing and reverse currrents.—S. K.

Magnetic Methods of Detecting Surface Flaws. P. Galan and I. D. Cenák. (*Zváranie*, 1957, **6**, (2), 34–38). [In Slovak]. The basic principles of magnetic flaw detection are reviewed, and a method of modifying a welding transformer to supply current for the magneto-inductive circuit of a simple testing device of this type is described. The performance of the instrument, and details relating to handling and accessories are discussed.—P. F.

Magnetic Domains in Evaporated Thin Films of Nickel-Iron. C. A. Fowler, jun., E. M. Fryer, and J. R. Stevens. (*Phys. Rev.*, 1956, **104**, Nov. 1, 645–649).

Studies on Anisotropic MK Permanent Magnet. I. On the Chemical Composition and the Additional Elements. T. Mishima and N. Makino. (*Tetsu to Hagane*, 1956, **42**, Nov., 1063–1066). [In Japanese]. Results are reported for the MK (Fe–Ni–Al) alloy. With increase of Ni up to 17%, coercive force is improved, and increase of Co up to 28% improves residual induction; the total of the two should not exceed 39%. The optimum Al content is 7·8–8·4%, and magnetic properties suffer below 7%. C content should be below 0·05%, but Ti additions overcome the deleterious effects. Si, Mn and P contents should be lower than 0·2%, 0·3%, and 0·02%, respectively. Cb and Zr improve the magnetic properties, while B, Cr, W, Mo, V, Be and Sn worsen them.—K. E. J.

Semi-columnar Magnets Give Improved Performance. (*Metalworking Prod.*, 1956, **100**, May 11, 522). Three new permanent magnet materials, semi-columnar in structure, are being introduced by members of the Permanent Magnet Association. These materials are modified forms of Columax with the advantage of being more easily adapted to existing methods of large-scale production.—M. A. K.

Electromagnetic Thickness-Meter. M. A. Berliner. (*Zavodskaya Laboratoriya*, 1956, **22**, (3), 342–344). [In Russian]. A device for measuring the thickness of non-metallic coatings on metal (ferromagnetic and non-magnetic) bases is described. The measurement is made from the reaction of an inductor coil on the eddy-current field.—S. K.

Rapid Method for the Analysis of Bessemer Steel by Measuring Thermal E.M.F. V. M. Yuferov, M. P. Kuznetsov, V. A. Nikitskaya, A. I. Novachok and I. I. Shargorodskiĭ. (*Zavodskaya Laboratoriya*, 1956, **22**, (4), 397–401). [In Russian]. The successful use of a method, based on measurements of thermal e.m.f. for control analyses of Bessemer steel for carbon and silicon is described. A massive copper bar serves as the hot electrode, the cold electrode being a water-filled iron box. The flat, quenched specimen is brought to a constant temperature of 20° C and placed between the electrodes, a temperature difference between those of 100° C being arranged. Carbon has a different effect on the thermal current in hardened and non-hardened specimens, but the effect of silicon remains unchanged: on the basis of this silicon can also be determined.—S. K.

Automatic Inspection in Works of Mild Steel Billets by Electromagnetic and Ultrasonic Methods. V. Husarek and L. Beaujard. (*Rev. Mét.*, 1957, **54**, Feb., 135–146). A review of these non-destructive test methods is first presented. The automatic application of the methods to the testing of billets for a modern wire mill, is discussed. The practical economic advantage of testing is discussed, with examples of operation with very sound billets and with billets containing defects.

The Application of Ultrasonics in Metallurgy and Mechanics. P. Hemardinquer. (*Rev. Gén. Méc.*, 1956, **40**, Dec., 429–433). Ultrasonic generation for use in the tinning of metals are described with examples of equipment available. The filling in of minor cavities on castings with a tin-zinc alloy using an ultrasonic soldering iron to insure perfect adhesion is described.

An Ultrasonic Impulse-Type Flaw Detector. J. Obraz. (*Zváranie*, 1957, 6, (3), 72–75). [In Slovak]. The design and possible applications of an instrument of recent Czechoslovak manufacture are discussed.—P. F.

Ultrasonic Attenuation in Steel and Cast Iron. K. Kamigaki. (*Sci. Rep. Tôhoku Univ.*, 1957, **9A**, Feb., 48–77). Pulsed longitudinal waves of 0·5–25 megacycles/sec were used on cast irons and steels with various textures. Cr–Mo and C steel were normalized, quenched, tempered to troostite-sorbite and attenuation measured at each stage. Austenite grain size was also investigated by heat treatment above the transformation point. White cast iron was tempered in stages to spheroidal graphite structure and flake graphite irons were also used. Results are summarized, the most important factor is Rayleigh scattering by grains and loss of energy depends upon size and anisotropy of the crystals.

New Tools Extend Uses for X-Rays. (*Iron Age*, 1956, **178**, Nov. 29, 87). Brief descriptions are given of two new radiographic tools which will extend the use of X-ray techniques in research and quality control. The X-ray microscope permits metallographic examination of the interior of specimens at magnifications up to × 1500, and also three-dimensional examination. The "xeroradiography" technique is a rapid, filmless method of producing X-ray images at low cost, using electrostatically-charged selenium-coated metal plate.—G. F.

Radiographic Inspection of the Structure of Lifting Chains. G. A. Homes and R. Pankowski. (*Metal Treatment and Drop Forging*, 1957, **24**, Apr., 143–148). A method of inspection for the detection and measurement of embrittlement in wrought iron and mild steel chains is described. Degree of cold work, sustained in service, is tested by back-reflection at the Bragg angle giving localized measurement, objective recording of intensity as a function of angle and degree of cold work by lattice distortion measurement. Recommendations for annealing and calculation of permitted load are indicated.

Some Characteristics of the Scintillation Counter and its Use for Technical Purposes. I. G. Fakidov and A. A. Samokhvalov. (*Zavodskaya Laboratoriya*, 1956, **22**, (6), 678–682). [In Russian]. An account is given of characteristics of scintillation counters which affect their suitability of gamma-ray detection with sufficient stability under works conditions. A possible arrangement of apparatus for this purpose is described.

Gamma-Ray Defectoscopy with a Scintillation Counter. I. G. Fakidov and A. A. Samokhvalov. (*Zavodskaya Laboratoriya*, 1956, **22**, (6), 673–677). [In Russian]. The use of scintillation counters in the detection of defects by γ-rays is discussed and the development of an instrument described. From experience in the use of this technique for testing steel parts up to 200–300 mm thick under works conditions it is concluded that it has several advantages over the photographic and the Geiger-Muller counter methods: high sensitivity and testing-speed, possibility of continuous testing of a moving object. The instruments developed are very stable and give reproducible results, the sensitivity being 2–2·5%.—S. K.

Hydrogen Embrittlement and Internal Friction of Carbon Steel. G. Mima and M. Mizuta. (*Tetsu to Hagane*, 1957, **43**, Feb., 132–137). [In Japanese]. Investigations were made on hypoeutectoid carbon steels. Although, macroscopically, diffusible H_2 is uniformly distributed at room temperature, it is microscopically more absorbed in free ferrite than in ferrite in pearlite. Its evolution follows Fick's law, and the diffusion coefficient is calculated. The recovery by ageing of the impact value is more rapid than that of internal friction, which itself is more rapid than that of ductility in tensile testing.—K. E. J.

The Influence of Hydrogen on the Mechanical Properties of Steel. P. V. Sklyuev, L. I. Kvator, and V. E. Sharipo. (*Stal'*, 1956, (10), 909–915). [In Russian]. The influence of H_2 on the static and impact strengths and plastic properties of three steels used for making critical parts for power equipment has been studied. The composition of the steels covered the range 0·32–0·35% C, 0·50–0·59% Mn, 0·26–0·32 Si, 0·92–1·20% Cr, 0·16–2·95% Ni, 0·26–0·36% Mo, 0·012–0·026% P, and 0·010–0·015% S. All specimens had a uniform sorbitic structure, H_2 being introduced during electrolytic etching. The modification of the effects of H_2 by storage of the steel, by temperature, by the strength level of the steel, by its composition (a further series of steels being used for this) and by its structure (for one steel) is discussed.—S. K.

Influence of Alloying Elements on the Content and Mobility of Hydrogen in Steel. Yu, A. Klyachko and T. A. Izmanova. (*Stal'*, 1957, (6), 507–511). Ti, Zr, C, and Mn tend to retain H_2 and others, such as Si, have no effect. Plastic deformation and heat treatment before deformation reduce H_2 content where the first group of elements are absent but have practically no effect in their presence. The amount of H_2 removed depends on the amount of metal, more is evolved, relatively, from small samples.

Studies on Cementite Cracks of High Carbon Steel. I. T. Uesugi. (*Tetsu to Hagane*, 1956, **42**, Dec., 1111–1117). [In Japanese]. Cavities occasionally observed at the centre of rolled high-C steel are attributed to very fine cracks (Acm cementite cracks), which grow during rolling. These cracks, of width less than 0·1 μ, are caused by supersaturated hydrogen at temperatures below 200° C in the Acm cementite of cooling billets after pressing or working. They can be observed as macro-hair-cracks after etching by HCl, but not with oxidizing etches or as flakes in fractured surfaces. II. (1957, **43**, Jan., 48–54). The effects of hydrogen additions to file, bearing, and tool steels were investigated. "Acm cementite cracks" are generated in hypereutectoid steels by additions of H_2, and supersaturated H_2 is the main cause of the cracks. The so-called macro-hair-cracks and flakes on the fractured surface are other manifestations of these cracks.—K. E. J.

Movement of Hydrogen in Solid Steel under the Influence of an Electrical Field. V. I. Yavoĭskiĭ and D. F. Chernega. (*Stal'*, 1956, (9), 790–793). [In Russian]. An account is given of experiments in which the effect of direct currents on the movement of H_2 in various steels and its desorption into the atmosphere were studied. High-C, Mn, medium-C, low-C rimming, Si, Si–Cr, Cr, and high-Cr steels were used, and for the first three a gradual movement of H_2 from the anode towards the cathode occurred, but the quantity of H_2 used was very small. No appreciable movement could be detected with the other steels, nor was there any appreciable effect of the current on the rate of desorption and removal of H_2 from the metal with any of the steels.—S. K.

The Central Laboratory of the "Bol'shevik" Chemical-Engineering Equipment Works in Kiev. N. S. Kozachenko. (*Zavodskaya Laboratoriya*, 1956, **22**, (10), 1261–1264). [In Russian]. The work of the various departments of the central laboratory at a factory producing chemical plant from a wide variety of stainless steels and other alloys is described.—S. K.

METALLOGRAPHY

Crystallographic Viewpoint and the Progress in Physical Metallurgy. (Honda Memorial Lecture). Z. Nishiyama. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Aug., M1–M11). [In Japanese]. The essential parts played by crystallographic and microscopical methods in the study of metals are described and illustrated by examples. Crystallographic methods using X-rays, electrons, and neutrons have provided final solutions to many problems in physical metallurgy, e.g. the nature of solid solutions, superlattices and precipitation, martensite formation and tempering in steels. (32 references).—K. E. J.

The "Electrolysenrohr," a New Apparatus for Metallographic Testing. L. G. Damgaard and E. Knuth-Winterfeldt. (*Metalloberfläche*, 1957, **11**, Mar., 75–76). This instrument consists of inner and outer tubes of stainless steel. The outer ends in a plastic sleeve, which is pressed against the specimen to be electrolytically polished. The electrolyte passes between the outer and inner tubes, and is withdrawn through the inner tube; the tubes are the cathode, and a D.C. of 1·5 A and 45 V is used. Examples of the use of the instrument are given.

Modern Metallographic Equipment. J. C. Wright. (*Metal Treatment Drop Forging*, 1957, **24**, May, 175–180; June, 249–253; July, 282–284). More advanced optical apparatus, photomicrographic and projection systems are illustrated and described, with a section on photographic emulsions, filters, and colour micrography.

Dislocations and Stacking Faults in Stainless Steel. M. J. Whelan, P. B. Hirsch, R. W. Horne, and W. Bollmann. (*Proc. Roy. Soc.*, 1957, **240**, July 16, 524–538). Direct observations of dislocations in deformed thin sections are made by electron microscopy.

Electron Microscope for Alloy Steel Research. (*Metallurgia*, 1957, **55**, May, 262). A brief description of a new Siemens

Mark II electron microscope installed at the Central Research and Development Dept. of the United Steel Co. Ltd. is given.

Electrolytic Polishing of Metals for Metallographic Observation. I. Hrivňák. (*Zváranie*, 1956, **5**, (2), 46–48). [In Slovak]. The preparation and uses of the electrolytic chromic oxide in acetic acid etch are discussed.—P. F.

Non-Metallic Inclusions in Steel. J. Skála and D. Tlustá. (*Hutník*, 1957, **7**, (2), 42–47). [In Czech]. The effects of inclusions on the mechanical properties of steels, their origin and identification are discussed. Precautions necessary in steelmaking in order to avoid undue amounts of inclusions are considered.—P. F.

Thermodynamic Investigation of Non-metallic Inclusions in Steel. II. Behaviour of Titanium, Zirconium and Vanadium Nitrides and Carbides, and their Effect on the Austenitic Grain Size of Steel. H. Sawamura and T. Mori. (*Tetsu to Hagane*, 1957, **43**, Jan., 31–38). [In Japanese]. The free-energy equations of the formation of the carbides and nitrides are derived, and the equilibrium relationships are calculated for various contents. Calculations show that titanium and zirconium carbide and nitride should be effective grain-growth inhibitors, but inclusions precipitated in austenite should be more effective than those in molten steel. VC does not precipitate in steel, but V_4C_3 may precipitate in austenite. (34 references).—K. E. J.

The Morphology of Carbide in α-Iron. W. Pitsch. (*Acta Met.*, 1957, **5**, Mar., 175–176). [In German]. Electron microscope studies show that the precipitation of carbide in carbonyl iron is in the form of platelets.—A. D. H.

The Shape of Carbide Precipitate Particles in α-Iron. R. H. Doremus. (*Acta Met.*, 1957, **5**, July, 393–397). Growth equations are derived for particles of various shapes and compared with internal friction data. It is concluded that for annealing temperatures over 100° carbide particles form as platelets spaced along the dislocation lines and become more nearly spherical as they grow.

Some Observations on Ferrite-Carbide Aggregates in Alloy Steels. E. S. Davenport. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, May, 677–688). The author reviews the development of austenite transformation studies in recent years, and discusses certain aspects of the subject. He refers particularly to the relationship between microstructure and physical properties of a Ni–Cr–Mo steel of eutectoid carbon content.—G. F.

Study of Carbides in Iron and Steel by Electrolytic Isolation. III. On the Graphitisation of Cementite in Cast Iron. T. Sato, H. Kaneko, and T. Nishizawa. (*Nippon Kinzoku Gakkai-Si* 1955, **19**, July, 445–448). Si, Ni, Cr, and Mn are soluble in cementite according to the distribution law. Isolated cementite graphitises by heating; the isothermal reaction rate is accelerated by Ni and Si, but retarded by Mn and Cr dissolved in solid cementite. The first stage of graphitization takes place not only by supersaturation of graphite into austenite but also by thermal decomposition of cementite. (11 references).

Study on Carbides in Practical Special Steels by Electrolytic Isolation. II. On Carbides in High-C, High-Cr Die Steel. T. Sato, Y. Honda, and T. Nishizawa. (*Tetsu to Hagane*, 1956, **42**, Dec., 1118–1122). [In Japanese]. Results are given for a die steel containing 2·25% C and 12·03% Cr, and comparisons are made with one containing 0·59% C and 4·73% Cr, representing the matrix. In the annealed state, carbides in the die steel had the formula $(Cr,Fe)_7C_3$, and were present in an amount of 26 wt.-%. With rise in quenching temperature, the carbides dissolve into austenite. The carbide precipitated by tempering above 650° C is $(Cr,Fe)_7C_3$; that by tempering at 400–500° C is of the $(Fe,Cr)_3C$ type. The two types co-exist at 550–650° C.—K. E. J.

Grain Growth in Steels. Role of Aluminium Nitride.—Part I. A. B. Chatterjea and B. R. Nijhawan. (*Metal Treatment and Drop Forging*, 1957, **24**, Jan., 3–6; Feb., 54–60). Grain growth is briefly reviewed and views on its inhibition in steels are examined. (53 references). These views are then discussed and a value for residual Al near 0·05% optimal for grain-growth inhibition is confirmed, below and above this value Al is relatively inefficient. Hypotheses based on alumina-nuclei effects, residual aluminium, surface adsorption, and presence of sulphide do not appear to account for all the phenomena. On soaking in N_2, absorption increased with Al content and developed a characteristic gradation in grain size. Above 0·4% Al precipitation of an acicular constituent at the grain boundaries was observed, increasing with residual Al. X-Ray diffraction showed the needles to be AlN. Its solubility in austenite does not explain effects above optimum concentration but mode of distribution may account for them.

The Influence of Impurities on the Recrystallization Texture of Cold-Rolled 3% Silicon Iron. J. D. Fast. (*Philips Res. Rep.*, 1956, **11**, Dec., 490–491). A study of the production of oriented polycrystalline sheet is reported. Pure material does not orientate. Nitrogen promotes the effect and must be expelled later by heating in H_2.

On the Method for Revealing the α Grain Size in Steel. R. Kadowaki et al. (*Tetsu to Hagane*, 1956, **42**, Sept., 838–840). [In Japanese]. The text is illustrated by good photomicrographs showing: (*a*) α-grain boundaries of 0·25% C steel, heat-treated at 1000° C and furnace-cooled for 30 min; (*b*) the same in 1·12% C steel similarly heat-treated; and the α-grain boundaries in a spheroidal-structure ball-bearing steel.—K. E. J.

Study on Hardenable Valve Steels by Sigma Phase Formation. Y. Matsunaga et al. (*Tetsu to Hagane*, 1956, **42**, Sept., 844–846). [In Japanese]. For several steels (compositions given) graphs are given of hardness against ageing temperature (500–900° C) for 15 h; hardness against ageing time (0–50 h) at 780° C; and mechanical properties against temperature (500–900° C).—K. E. J.

On the Relations between Grain Size and Mechanical Properties of High-Manganese Steel. C. Muto et al. (*Tetsu to Hagane*, 1956, **42**, Sept., 873–874). [In Japanese]. The effects of mould diameter and thickness on grain size are discussed, as is the effect of heating times up to 5 hours after casting at 1630°, 1580°, and 1520° C. For casting temperatures of 1600°, 1550°, and 1500° C, grain sizes and tensile properties are tabulated.

Simple Orientation Relationships For Secondary Recrystallisation in Si–Fe. C. G. Dunn and P. K. Koh. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, Jan., Section 2, 81–86). The authors present data on the textures after primary and after secondary recrystallizations which can be understood from the point of view of the oriented nucleation growth selectivity theory. A simple direct relationship is shown connecting the cold-rolled texture and components in the secondary recrystallization texture. An explanation is given for large primaries (nuclei) occurring in specific orientations.—G. F.

Pressure-Induced Diffusion and Deformation during Precipitation, Especially Graphitization. M. Hillert. (*Jernkontorets Ann.*, 1957, **141**, (2), 67–89). Flow by diffusion in non-isobaric systems is related to activity gradient and the pressure dependence is derived. Differences in volume of precipitate and matrix are derived and formulæ given for mechanical deformation due to the resultant pressure and for deformation energy. Elastic, plastic, and creep models are all used. The creep model applied to graphite growth in steel and the effects of Mo and Cr gives good agreement with experiment. It is suggested that in cast iron carbon diffusion is the rate-controlling process when the particles are large, when small deformation of the austenite by creep may play a part.

The Different Forms of Graphite Found in Unalloyed Grey Cast Iron. M. Ferry. (*Fonderie*, 1956, Oct., 395–415). Following a brief description of the A.S.T.M. classification system for graphite, the mechanism of solidification of grey cast iron and the factors affecting the distribution, form, and size of the graphite flakes are discussed. Detailed metallographic results are then presented and discussed to show the effects of cooling rate, superheating and casting temperature, chemical composition, and ladle additions on the graphite structure. In conclusion the factors favouring the development of each type of graphite in the A.S.T.M. classification are summarized.—B. C. W.

Investigation on Cast Iron Having Fine Graphites Produced by Melting Cast Iron Covered with Slag Containing TiO_2. VIII. Influence of Addition of Metallic Titanium, Ferro-titanium and Copper-titanium Alloys upon the Structure of Grey Cast Iron. M. Tsuda et al. (*Tetsu to Hagane*, 1956, **42**, Sept., 809–811). [In Japanese].—K. E. J.

Fundamental Studies on Spheroidal Graphite Cast Iron. XVI. Influences of Contents of C and Si on Formation of Spheroidal Graphite. R. Ozaki et al. (*Tetsu to Hagane*, 1956, **42**, Sept., 805–807). [In Japanese].—K. E. J.

Study on Graphitic Steel. I. Effect of Silicon on Graphitic Steel. T. Sato and S. Toya. (*Nippon Kinzoku Gakkai-Si*, 1956, **20**, Jan., 5–9). [In Japanese]. Graphitization conditions and properties were determined for "Graph-Sil" steel con-

taining 1·1–1·5% C and 0·7–1·5% Si. The silicon content is the main factor controlling the precipitation of free graphite, and it must be held within definite limits for forging. The suitable graphite content is about 0·6%.—K. E. J.

The Influence of Tellurium and Lead on the Occurrence of Some Abnormal Formations of Graphite in Grey Cast Irons and Their Effects on the Mechanical Properties. G. E. Morton. (*Brit. C.I. Res. Assoc.*, 1956, **6**, Dec., 436–443). An examination is first reported of 5 commercial castings which failed before or during service and indicated that failure was due to abnormal forms of graphite. The effect of tellurium (derived from tellurium-bearing mould or core washes or as an addition of metal) and lead (derived from lead-bearing materials in the charge) on the structure has been studied and the structures found defective in practice have been reproduced in the laboratory.—B. G. B.

Rate of Grain Growth of Austenite. Y. Masuko. (*Tetsu to Hagane*, 1956, **42**, Sept., 784–785). [In Japanese]. For three steels (compositions given), having fine-, coarse-, and duplex-grained structures, the effects on grain size of austenitizing for 6 hours at temperatures between 850° and 1200° C, and at 925° C for times between 30 min and 10 hours, are shown graphically.—K. E. J.

The Effect of Alloying Elements on Austenite Grain Size by a Thermal Etching Method. III. H. Hirotani et al. (*Tetsu to Hagane*, 1956, **42**, Sept., 785–787). [In Japanese]. The investigations include the effects of Al, Zr, N, Nb, Cu, and Ti.—K. E. J.

A Study on the Method for Revealing the Austenite Grain Size in Steel. K. Mizukawa et al. (*Tetsu to Hagane*, 1956, **42**, Sept., 787–788). [In Japanese]. A copper diffusion method is described, and used to estimate the effects of temperature and time on the grain size of steel.—K. E. J.

Studies on Ball Bearing Steels. IX. Austenitizing Behaviour and Rapid Life Test of Drawn Ball Wire. M. Ueno et al. (*Tetsu to Hagane*, 1956, **42**, Sept., 763–765). [In Japanese]. Results are given, for steels of quoted composition and charge make-up, of the amount and size of non-metallic inclusions, and the relationships between quenching temperature and bending stress and retained cementite content.—K. E. J.

On the Kinetics of the Pearlite Reaction. J. W. Cahn. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, Jan., Section 2, 140–144). The author re-examines existing kinetic data on the pearlite reaction in terms of recently derived rate laws. The nucleation rate varies more rapidly with temperature than previously supposed, and with eutectoid steels below 660° C this rate is high enough for the transformation rate to be independent of it. An analysis of the pearlite reaction in the presence of carbide particles supports the two-stage nucleation concept.—G. F.

Recent Progress in the Study of the Structure of Martensite and its Decomposition Products in Carbon Steels. K. Kuo. (*Jernkontorets Ann.*, 1956, **140**, (11), 854–904). A very extensive review is given based largely on Russian work which is treated in full so that reference to originals should be unnecessary. The work is mainly based on X-ray studies but is supported by other evidence. The structure of martensite is given with an equation relating parameters to carbon content. Particle size is about 200 Å giving great line broadening except when isolated with removal of stress, which is thus shown to be non-uniform, elastic, and inter-martensitic. Displacement of iron atoms is larger the nearer they lie to carbon and can be visualized as frozen thermal vibrations, but there is also some isotropic dynamic lattice distortion weakening Fe–Fe bonds. Tempering proceeds by precipitation of ϵ-carbide and formation of low-C martensite, transformation of retained austenite to bainite and cementite precipitation. Structure and composition of ϵ-carbide are described, and it appears in acicular form regularly oriented with respect to martensite. The rejection of carbon is discontinuous and non-homogeneous and low-carbon martensite forms with a modified first-order reaction rate. Cementite precipitated below 400° C has a platelet habit of growth but above 600° normal sharp X-ray patterns appear. Hägg carbide is probably absent. As tempering increases the ferrite particle size increases and internal stress is reduced, at 450° C it becomes negligible. (128 references).

A Contribution to the Thermodynamics of Graphitization. S. Drapal. (*Slevarenstvi*, 1957, **5**, Mar., 75–80). The free energy of cementite formation is calculated and its instability in the solid phase is shown, and the difference of free energy between stable and metastable states of the Fe–C system is related to C concentration.

Metallography of a Medium Carbon Steel Subjected to Slow and Impact Compression. J. D. Campbell, J. Duby, and K. E. Puttick. (*Phil. Mag.*, 1957, **2**, 8th Series, Apr., 548–553). Optical and electron microscopes were used. In slow deformation, coarse slip occurs in narrow regions on corrugated surfaces, apparently due to the presence of an ageing precipitate. Coarse slip is absent in rapidly strained steel, fine slip occurs with occasional formation of deformation bands. Twins were only observed once in a specimen of abnormal structure.

On the Determination of Austenite Grain Size of Steel by the Oxidation Method. Y. Masuko and T. Onitake. (*Sumitomo Metals*, 1956, **1**, Oct., 36–54). [In Japanese]. Many steels are examined by oxidation and other methods and by heat etching, in which it was confirmed that oxidation has an inhibiting effect on grain growth. The oxidation method gives the same results as other processes and is very suitable for industrial use. (54 micrographs, 14 references).

Structure and Magnetic Properties of Permanent Magnet Alloys During Isothermal Precipitation Hardening. E. Biederman and E. Kneller. (*Z. Metallkunde*, 1956, **47**, May, 289–301; Dec., 760–774). Section I of this paper deals with particle growth, critical particle size and coercive strength; the process of particle growth during isothermal disintegration of several Cu–Ni–Fe alloys was determined as a function of annealing time. Electron microscope investigations confirmed the correctness of the method according to Daniel and Lipson for determining the length of periodicity of fluctuations in concentration, and verified these authors' conclusions. The maximum coercive forces observed during hardening are explained by the anisotropy of the ferromagnetic structural constituents. In the second section, other properties are measured in relation to ageing time, and the general principles are given for the magnetic behaviour of fine ferromagnetic particles in a non-magnetic matrix. It is concluded that the alloy is almost completely segregated during quenching into coherent Cu and Fe–Ni rich cluster-like regions with an average diameter of 50–100 Å.—L. D. H.

Zone Melting. W. G. Pfann. (*Mrt. Rev.*, 1957, **2**, (2), 29–76). A review of the theory with a very brief note on iron. (79 references.)

Polymorphism of Metals and the Influence on it of Alloying Elements. W. Koster. (*Met. Ital.*, 1957, **49**, Jan., 14–23). [In Italian]. The author discusses polymorphism, i.e. the phenomenon by which an element can exist in different crystalline forms. The kinetics of polymorphism and the mechanics of the transformations are studied. The influences of additions of alloying elements on polymorphism are discussed with special reference to ferrous systems and Co and Ti alloys. The polymorphism of Mn is also described. (16 references.)

Decomposition of Delta-Ferrite in High-Alloy Steels. K. Kuo. (*Jernkontorets Ann.*, 1957, **141**, (3), 146–174). In ternary systems of Fe, C, and V, Cr, Mo or W the δ-ferrite has a content of the alloying metal between those of austenite and the carbide (VC, $Cr_{23}C_6$, Fe_3MO_3C or Fe_3W_3C) so that the former can give rise to the latter. This aggregate resembles pearlite if formed above 850° C and the term δ-eutectoid is applied to it. Below 850° C it may be acicular, resembling bainite. Its formation is governed by short range diffusion of the alloy element and occurs at temperatures higher than those for pearlite formation but kinetics and morphology are similar. As both transformation products have increased C content the eutectoid formation is soon inhibited by carbon deficiency but in presence of coexisting austenite C may diffuse and take the change to completion. 18–4–1 High-speed steel consists of austenite and Fe_3W_3C at 1300° C, above this δ-ferrite appears first in low-C and liquid first in high-C steels, the former decomposing on cooling to a dark-etching δ-eutectoid with a carbide border. In low-C 18–10–3 Cr–Ni–Mo steel decomposition starts with austenite formation often entrapping unchanged ferrite. Sigma-phase forms later mainly in δ-ferrite not yet transformed. (32 references).

On the Existence of a Plastic Deformation in Iron during the $\alpha \rightleftharpoons \gamma$ Transformation. P. Lehr. (*Compt. Rend.*, 1956, **244**, Jan. 2, 77–80). A metallographic study shows distinct flow during transformation while hot.

Investigations of the Isothermal Time–Temperature–Transformation Diagrams and the Quench-Hardenability of Steels. E. Greulich. (*Arch. Eisenhüttenwesen*, 1957, **28**, Feb.,

91–100). These diagrams were determined for six steels having 0·16 to 0·72% C, 0·2 to 1·2% Si, 0·6 to 1·4% Mn, 0 to 1·6% Cr, 0 or 0·36% Mo, 0 or 1·5% Ni, and 0 to 0·27% V. It was found that, in the gap between pearlite and intermediate transformation, there is no stable austenite, but that transformation is inhibited. The hardenability index was formulated as the natural logarithm of the shortest half-life period of transformation, which enabled hardenability to be calculated for different steels on a new basis.—L. J. L.

On the Continuous Cooling Transformation Curve of High Speed Steel. I. S. Owaku and H. Akasu. (*Nippon Kinzoku Gakkai-Si*, 1954, **18,** June, 341–344). The curve for 18–4–1 steel JES SKH 2 was determined by a process described. Austenite is most unstable at ~ 720° and the critical cooling rate is ~ 17°/min.

The Present State of Knowledge on the Austenitic Transformation of Steel. A. Rose. (*Schweiz. Arch. Wiss. Techn.*, 1956, **22,** Oct., 305–318). A general discussion of kinetics of the austenite transformation is given. Illustrations of typical transformation–temperature–time curves are shown for isothermal and continuous cooling conditions for a wide range of steels. The relation between these diagrams and heat treatment is discussed.—J. R. P.

A Survey of the Effect of Austenizing Temperature and Rate of Continuous Cooling on the Structure and 700°–1200° F Properties of Three Low-alloyed Steels. K. P. Mackay, A. P. Coldren, A. I. Rush and J. W. Freeman. (*PB* 121149, 1956, pp. 76; *TIDU List* 723). Types Ni–Cr–Mo, 1·25 Cr–Mo–V and 3 Cr–Mo–W–V were studied on 1-in. rounds oil quenched and 3-in. and 6-in. simulated rounds, air-cooled. Microstructures were examined and best combinations of strength and ductility were obtained by normalizing from the lower austenizing temperature.

Metallographic Study of Structural Changes During Tempering of High Carbon Steel. L. Bezděk and D. Růžička. (*Problems and Perspectives of Czechoslovak Metallurgy and Foundry*, 1956, 90–103). A systematic optical and electron micrographic study of quenched 1·42% C steel tempered in the range 50°–700° C is reported. Dilatometric studies, microhardness of martensite and residual austenite and their relation to tempering conditions are reported and correlated with structure.

Carbide Precipitation in Corrosion-Resistant Titanium Stabilized Steels Type 18–8. V. Cihal and P. Gröbner. (*Problems and Perspectives of Czechoslovak Metallurgy and Foundry* 1956, 104–115). The kinetics of carbide precipitations are evaluated in regard to a study of intercrystalline corrosion. High homogenization temperatures partially destroy the TiC and precipitation takes place between 550–850° C. TiC precipitates at higher temperatures. Intercrystalline corrosion appears in presence of $Cr_{23}C_6$ apparently due to the higher diffusion rate of Cr at 750° compared with Ti. Isolated carbides were examined with X-rays, confirming the theory.

The Dimensional Stability of Ball-bearing Steel. III. T. Hattori. (*Tetsu to Hagane*, 1956, **42,** Nov., 1053–1059). [In Japanese]. X-Ray diffraction methods were used to study the effects on amount of retained austenite of quenching temperatures, tempering after quenching, air cooling between quenching and tempering, secondary tempering, subzero treatment, marquenching, and the mass of the specimen; and the reduction of spheroidised carbide by higher quenching temperatures and repeated quenching and its diffusion into the matrix.—K. E. J.

On the Properties due to Heat-treatment of [Cr–Mo–V] Die Steel for Hot Work. S. Koshiba, K. Tanaka and A. Inada. (*Tetsu to Hagane*, 1956, **42,** Nov., 1060–1063). [In Japanese]. For several steels in the series, transformation points, hardness and rate of deformation after various heat-treatments, and changes in hardness and impact value at high temperatures were studied. The steels were compared with W–Cr–V steels for drawing of non-ferrous shapes, and optimum heat-treatments were ascertained.—K. E. J.

Contribution to the Phase Analysis of Steels. B. Šestak. (*Problems and Perspectives of Czechoslovak Metallurgy and Foundry*, 1956, 125–141). Quantitative determination of γ-phase and investigation of the kinetics of austenite transformation by magnetic methods are described. The effects of field intensity and test piece shape are given. Apparatus is described and used to determine the relation of the magnetisation intensity of martensite to carbon content and to the austenite transformation in eutectoid steels. (18 references).

On the Structures of Some Chrome–Nickel Steels. J. Bourrat, L. Colombier, J. Hochmann and J. Philibert. (*Compt. Rend.*, 1957, **244,** Feb. 25, 1197–1200). Steel with 0·27% C, 4·9% Ni and 25% Cr shows islets of austenite in ferrite on heat-tinting in air. The ferrite contains 24·5% Cr, 4·5% Ni and the austenite 21·5% Cr, 6·5% Ni. Other equilibria are possible.

The Character of the Peritectic Reaction of the Binary Fe–C Systems within the Transitions of the Ternary Fe–C–P System. P. H. Tütsch. (*Schweiz. Arch. Wiss. Techn.*, 1956, **22,** Nov., 345–355). A study of the peritectic transformation in the Fe–C–P system in which the theoretical possibilities are outlined and the results of experiments described. It is shown that of the two possible γ phases existing at temperature below the peritectic temperature (993°–998° C) the γ phase cannot be suppressed by quenching. Quenching in water from temperatures above the peritectic temperature prevented the formation of crystals in the γ_2 phase. Illustrations of microstructure are given.—J. R. P.

Microporosity in Steel. (*Castings*, 1957, **3,** May, 25, 27). Interdendritic porosity is briefly considered and vacuum melting has been found to minimize it.

Some Methods of Determining Recrystallization Temperatures. G. K. L'vov. (*Zavodskaya Laboratoriya*, 1956, **22,** (10), 1198–1200). [In Russian]. Based on comparative experiments with low-carbon alloys of iron with carbon and silicon a simple procedure for determining recrystallization temperatures has been developed. From the data obtained the activation energies for recrystallization have been found, taking that for silicon-free ferrite as unity, that for 3·7 atom % Si is 1·8 and for 6·9 atom % Si it is 3·2.—S. K.

Habit Planes of Martensite In Carbon-Chrome Steel. H. M. Otte and T. A. Read. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9,** Apr., Section 2, 412–417). From observation of the scatter of the habit planes of martensite in a 2·8% Cr, 1·5% C steel, the authors deduce that most of the scatter is real and only a small amount due to experimental error. Possible reasons for the real scatter are discussed.

Unique Structure of Pearlite Deformed by Explosive Load. E. W. La Rocca. (*Acta Met.*, 1957, **5,** July, 408–410). A stepped structure is illustrated and discussed.

The Mechanism of Formation of Banded Structures. P. G. Bastien. (*J. Iron Steel Inst.*, 1957, **187,** Dec., 281–291). [This issue].

Dislocation Energy as a Driving Force for Boundary Migration. C. G. Dunn and K. T. Aust. (*Acta Met.*, 1957, **5,** July, 368–370). Observations on a 3·25% Si iron cold-rolled to 2·5% and annealed at 1200° C are described in which grain boundary movement was measured. The difference in free energy across the boundary was estimated at 42,000 ergs/cm³ and the energy of a single dislocation at 14×10^{-4} ergs/cm. Possibilities of more refined measurements are indicated.

Structural Changes Associated with Strain-Induced Grain Boundary Migration in Si–Fe. K. T. Aust and C. G. Dunn. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9,** Apr., Section 2, 472–478). Using a chromic–acetic acid electro-etching method, the authors have studied the dislocation configurations existing in bicrystals of Si–Fe before and after strain-induced grain boundary migration. Structural observations are presented, and the variation of dislocation density in the region swept out by boundary migration is discussed.—G. F.

The Effect of Aluminium Nitride Precipitation in Relation to Grain Growth in Steels. A. B. Chatterjee and B. R. Nijhawan. (*Trans. Indian Inst. Met.*, 1954–55, **8,** 225–247, discussion 248–251). Grain growth inhibition in steel is discussed and an investigation using metallographic, X-ray, and chemical methods confirmed that precipitation and re-solution of fine aluminium nitride particles and their agglomeration in high-Al steels could account for the observed grain coarsening characteristics.—A. D. H.

Crystal Growth in Chemical Displacement. A. Goswami. (*J. Sci. Ind. Res.*, 1957, **16B,** May, 186–189). Electron diffraction studies of deposits obtained by displacement on Fe(100) faces and other metals are described and compared with cathodic deposits.

Dilatometer for Investigating Phase Transformation in Alloys. V. G. Permyakov and M. V. Belous. (*Zavodskaya Laboratoriya*, 1956, **22,** (10), 1251–1252). [In Russian]. A simple optical-mechanical dilatometer and its use with steel

specimens are described. The instrument is distinguished by high magnification and steadiness of readings.—S. K.

Fundamentals of the Transition Temperature Phenomenon in Steel. N. Gensamer, J. O. Brittain, L. O. Chang, and H. Hahn. (*PB* 117792, 1954, Aug., *TIDU List* 751, pp. 7). A ductile iron was prepared and the effect of hydrogen on ductile-to-brittle transition measured. Solid state outgassing, internal friction, and strain-ageing are considered.

Method for Studying the Isothermal Decomposition of Structurally Free Carbides in Cast Iron. M. V. Voloshchenko. (*Zavodskaya Laboratoriya*, 1956, **22**, (10), 1194–1197). [In Russian]. The study of isothermal decomposition of structurally free carbides in iron, based on measurements of the volume changes of a specimen with the aid of an optical differential dilatometer, is described. The best standard was found to be a specimen identical in form and chemical composition to the test piece, but previously annealed. Chromium plating of both was shown to be necessary to prevent decarburization. A carbide-decomposition kinetic curve and some microstructures obtained at various stages are presented, good agreement being evident.—S. K.

Diagram of the Eutectoid Transformation of Austenite in Grey and Malleable Cast Iron. K. P. Bunin, N. G. Osada, and S. A. Fedorova. (*Liteĭnoe Proizvodstvo*, 1956, (4), 17–19). [In Russian]. Existing austenite eutectoid transformation diagrams are criticised and a modified form is proposed for grey and malleable cast irons. These diagrams can be used not only for analysing isothermal phase transformations but also for those occurring during continuous cooling; from these diagrams and cooling curves structural changes can be forecast.—S. K.

Studies on Ball-bearing Steels. V. Solution of the Carbide into Austenite. M. Ueno and Y. Nakano. (*Tetsu to Hagane*, 1957, **43**, Jan., 39–43). [In Japanese]. At each heating temperature, hardness increases with holding time. At 840° C it increases rapidly and reaches a maximum after 1 h, while at 860° C it reaches a high value after only 5 min. Comparisons are made of the behaviour of carbide in austenite, for SKF and Japanese steels, and some observations are made on fracture roughness.—K. E. J.

Effect of Mo and V Additions on Si–Cr–W Steels for Shock-resisting Tools. S. Koshiba and T. Kuno. (*Tetsu to Hagane*, 1957, **43**, Feb., 142–145). [In Japanese]. Experiments were made on shock-resisting steel containing 0·55% C, 1·4% Cr, and 2·1% W, in which the transformation-point, quenching and tempering hardness, hardenability, rate of deformation, mechanical properties at elevated temperature and microstructure were investigated. Additions of 0·5% Mo and 0·3% V markedly improved the properties of the samples, which contained 0·26% or 1·03% Si.—K. E. J.

Thermodynamic Properties of CS and Solutions of Sulfur in Carbon-Saturated Liquid Iron. C. J. B. Fincham and R. A. Bergman. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, May, 690–694). The authors have studied the equilibria, at 1600° and 1800° C, between dilute solutions of S in carbon-saturated iron and gas mixtures consisting initially of A and CS_2. Comparison of the results with those obtained by previous workers with H_2–H_2S systems permits evaluation of the free energy of formation of CS and the determination of the equilibrium sulphur pressures in the systems.—G. F.

New Intermediate Phases in the Iron–Niobium System. H. J. Goldschmidt. (*Research*, 1957, **10**, July, 289–291). A high-temperature σ-phase, a η-carbide with 60% Nb, a Laves phase with an extensive homogeneity range (22–42 at.%), significant terminal solubilities and possibly a fourth compound very rich in Nb are indicated and a tentative revised diagram given.

CORROSION

A Corrosion Problem in Large Steam Generating Stations. T. J. Finnegan. (*Corrosion*, 1957, **13**, June, 405–409). Internal boiler-tube corrosion is discussed. Pinhole leaks appear and gradually enlarge, heavy deposits of black oxide often occur near the point of failure. Protective measures are considered such as " free caustic " control, acid cleaning, and the use of amine inhibitors.

The Blackening of the Interior of Preserve Tins, Its Metallurgical Causes, and Methods of Prevention. J. Teindl. (*Prümysl Potravin*, 1956, **8**, (2), 68–73). [In Czech]. Recent work, including the author's own researches, are reviewed. Protection of tins by means of the Protecta-Tin process is recommended.—P. F.

Underground Corrosion. M. Romanoff. (*Nat. Bur. Standards Circular* 579, 1957, Apr., pp. 227). A review of the work of the Bureau in the last 45 years is presented. Soils, corrosion theory and mechanism underground, types of corrosion, field tests on wrought and cast iron and steel, low and high alloy steels and high-Si and malleable irons, factors in ferrous corrosion, studies on bolts in the laboratory and the field, non-ferrous metals and non-metals, behaviour of metallic coatings, Zn, Pb, Al, organic coatings, porcelain enamel, laboratory, accelerated and field tests, soil corrosivity studies, stray current effects, measurement of corrosion rate, cathodic protection in theory and practice, circuits and methods are all surveyed. (407 references).

Corrosion Failure of Welded Joints of 18/8 Stainless Steel Along the Line of the Weld. B. I. Medovar and N. A. Langer. (*Avtomaticheskaya Svarka*, 1955, (6), 85–95). [In Russian]. " Knife line " corrosion along the heat-affected zone (fusion line) of an 18/8 weld is a type of intergranular corrosion which can occur in stabilized 18/8 steel both in oxidizing (e.g. HNO_3) and in non-oxidizing (e.g. H_2SO_4) media. The main cause is the dissolving of titanium carbides or Nb–Ta in austenite due to the heating up of the base metal to above 1300° C and the subsequent precipitation of the chromium carbides at the austenite grain boundaries. The precipitation of the carbides and the chromium impoverishment of the boundary portions of austenite occur either with slow cooling of the weld or with repeated heating of the bead zone up to about 650° C. To avoid this type of corrosion, the amount of stabiliser should be increased, a two-phase (austenite and ferrite) steel with a fibrous structure of primary ferrite should be used, care should be taken to avoid dangerous heating of the welded joint when laying down other beads in the vicinity and the weld should be diffusion annealed.—R. S.

On the Mechanism of Intercrystalline Corrosion of Mild Steel by Ammonium Nitrate. M. Smialowski and T. O. Ostrowska. (*Corrosion et Anti-Corrosion*, 1957, **5**, Mar., 76–81). Cathodic polarization tests on fine and coarse grained iron specimens and single crystals are reported and the changes appearing under tension noted. It is concluded that the crystallographic orientation of the grains plays the principal part in the phenomenon. Oxygen diffusion and the effects of fatty acids are also considered and the effect of tension on the cracks formed aids in the break-up of the metal.

Theoretical Studies and Laboratory Techniques in Sea Water Corrosion Testing Evaluation. F. L. LaQue. (*Corrosion*, 1957, **13**, May, 303–314). Apparatus for immersion tests is described and high velocity and turbulence and jet impregnated tests are also reported. Apparatus for corrosion fatigue testing in sea water is shown and crevice corrosion, galvanic potentials, polarization and cavitation erosion are mentioned. Sodium chloride solutions have advantages in the laboratory but sea water produces rather different effects and a synthetic sea water solution is devised.

The Corrosion of Steel in a Reinforced Concrete Bridge. R. F. Stratfull. (*Corrosion*, 1957, **13**, Mar., 173–178). Excessive spalling of concrete from the steel reinforcement resulted from rapid corrosion of the steel in a salt water environment. Electrochemical corrosion on a macro-scale was involved. Resistivity and corrosion cell data are presented and related to the deterioration of the concrete. Cathodic protection offers a possible remedy but requires certain precautions. In new work the use of concrete of minimum porosity is recommended together with a coverage of concrete over the reinforcement steel in excess of the 2-in. cover found inadequate in the present instance.—J. F. S.

Corrosion of Automobile Bodies. (*Tek. Tidskr.*, 1957, **87**, Mar. 26, 299–330). [In Swedish]. Non-protective corrosion film formation takes place in protected positions in a car body, where the metal does not have the opportunity to dry quickly (cf. *S.A.E. Trans.*, 1956, **64**, 221–26). In these circumstances high-carbon steel offers good resistance to corrosion, whereas Cr steel is prone to attack. Non-protective rust consists mainly of Fe_3O_4, whilst protective films are built up principally from Fe_2O_3.—L. D. H.

Casing Corrosion in the Petroleum Industry. J. L. Battle. (*Corrosion*, 1957, **13**, 132t–138t). Causes and means of prevention are listed. New methods of investigation such as the use of casing potential logs and current withdrawal-potential

surveys are described and the use of cathodic protection is discussed.

Corrosion by High Purity Water. Introduction to Symposium on Corrosion by High Purity Water. J. F. Eckel. **(i) Corrosion of Structural Materials in High Purity Water.** A. H. Roebuck, C. R. Breden and S. Greeberg. **(ii) Corrosion Engineering Problems in High Purity Water.** D. J. DePaul. **(iii) The Importance of High Purity Water Data to Industrial Applications.** W. Z. Friend. (*Corrosion*, 1957, **13**, Jan., 69–85). Papers presented to a symposium outline the aims of N.A.C.E. Unit Committee T–3–F organizing it and then separately discuss the following aspects: (i) the general corrosion resistance of classes of metals and alloys in closed-loop corrosion tests and the effects of surface preparation, oxygen concentration and other factors. In high purity water at elevated temperatures the highest corrosion resistance was shown by austenitic stainless steels, precipitation hardening stainless steels, Co alloys, Pt, Ti, Zr and Hf. Ferritic and martensitic stainless steels were not so good and plain carbon steels had poor resistance. (ii) Crevice, galvanic, intergranular and stress corrosion as they arise in systems recirculating high purity water. The dangers inherent in corrosion product build-up where tolerances are close are mentioned together with examples of pitting troubles on rivet heads. Temperature and oxygen concentration data are presented. 18–8 Stainless steels, hard chromium plate and straight chromium stainless steels were included in the experience reported. (iii) Modern industrial uses of high purity water. Boiler feed water is the greatest single use but many other industries handle high purity water. Their experience indicates the high corrosivity of high purity water is best controlled by stainless steels, nickel alloys or aluminium at high temperatures or by rubber or vinyl polymers at moderate temperatures.—J. F. S.

The Microdetermination of Dissolved Oxygen in Water. I. The Nature of the Problem. II. The Design of Water Sampling Vessels. E. C. Potter. **III. Titrimetric Determination of Iodine in Sub-microgramme Amounts. IV. Test of Winkler's Reaction to below 0·001 p.p.m. of Dissolved Oxygen.** E. C. Potter and J. F. White. (*J. Appl. Chem.*, 1957, **7**, June, 285–297; 297–308; 309–317; 317–328).

Corrosion Studies in High Temperature Water by a Hydrogen Effusion Method. M. C. Bloom, M. Krulfeld, W. A. Fraser and P. N. Vlannes. (*Corrosion*, 1957, **13**, May, 297–302). Measurements of the corrosion rate of low-C steel in distilled water at 600° F are reported and the method used is described. The effects of rising pH, thermal shock, temperature rise, heat treatment, and comparisons with stainless steel are given and the effect of wall thickness on hydrogen effusion at 600° F is measured.

The Effect of Dissolved Oxygen on the Corrosion of Steel and on the Current Required for Cathodic Protection. E. Schaschl and G. A. Marsh. (*Corrosion*, 1957, **13**, Apr., 243–251). The interrelationship of corrosion rate of steel, dissolved oxygen concentration, agitation, electrode potential and cathodic protection current density were studied under controlled laboratory conditions. The corrosion rate of steel and dissolved O_2 concentration are linearly related, the rate falling to zero at 1 p.p.m. Agitation largely affects the rate above 2 p.p.m. but has no effect below 2 p.p.m. dissolved oxygen. It is suggested that by measuring in the field the corrosion rate of an unprotected specimen it should be possible to compute the minimum current density needed for cathodic protection. In acidic, or corrosive neutral, environments it appears that iron atoms are lost by undermining at the anode so that less current is required for protection then corrosion rates suggest. The effect of differential aeration on corrosion rates is discussed in detail.—J. F. S.

Effect of Hot Hydrogen Sulphide Environments on Various Metals. F. J. Bruns. (*Corrosion*, 1957, **13**, Jan., 27–36). Catalyst beds for upgrading naphtha and for S removal suffer clogging by FeS scale. Plant and bench corrosion tests suggest that at 750–950° F low-chromium steels are only suitable if the H_2S concentration is below 100 p.p.m. At greater concentrations over 13% Cr steels are required. Calorized, hot-dipped, and hot-diffused coatings on carbon steel gave good service, the calorized being best.—J. F. S.

High Temperature Sulphide Corrosion in Catalytic Reforming of Light Naphthas. C. Phillips, jun. (*Corrosion*, 1957, **13**, Jan., 37–42). The problem of high temperature sulphide corrosion in catalytic reforming plants leading to plugging of the catalyst with FeS scale is discussed. Corrosion rates were investigated using steel shims inserted in by-passes on the plant and measuring their change in electrical resistance. Up to 12% Cr conferred improved resistance on carbon steels. In a pilot-plant, scaling began on all steels tested at 0·0075 to 0·008 mol% H_2S at 1000° F. Aluminized steel showed good resistance except at cut edges which peeled.—J. F. S.

High-Temperature Corrosion of Alloys Exposed in the Superheater of an Oil-Fired Boiler. D. W. McDowell jun., R. J. Raudebaugh and W. E. Somers. (*Trans. Amer. Soc. Mech. Eng.*, 1957, **79**, Feb., 319–328). Test racks holding ~ 30 samples were exposed in the gas inlet of the second bank of a boiler superheater. Scale and ash were analysed and corrosion rates measured as penetration in in. during test and calculated to in./year. All samples were examined metallographically. Various alloys including plain carbon and a few alloy steels were used. Attack by vanadates was observed.

An Investigation into the Catastrophic Corrosion of a Flake Graphite Cast Iron in Carbon Dioxide Containing a Small Amount of Carbon Monoxide. S. R. Billington and B. C. Woodfine. (*Metallurgia*, 1957, **55**, May, 213–220). A pearlite flake graphite cast iron has been found to undergo severe oxidation in times of less than 3000 hours in a CO–CO_2 equilibrium mixture under a pressure of 200 p.s.i. at 525° C. The oxidation behaviour of the iron in both CO_2 and CO–CO_2 mixtures at temperatures up to 525° C is described and a mechanism is suggested to account for the observed phenomena.—B. G. B.

Corrosion of Metals in Alkaline Solutions. VI. The Anodic Behaviour of Metals and the Effect of Oxidisers. H. Endo and N. Ohtani. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Aug., 502–504). [In Japanese]. The anodic behaviour of iron and other metals was investigated in solutions of NaOH and Na_2CO_3. The cathodic reaction was shown to be an oxygen electrode reaction. The difference in the electrode reaction in H_2O_2 and $K_2Cr_2O_7$ is probably due to atomic oxygen and adsorption of CrO_4^{--}, respectively.—K. E. J.

A Study of the Corrosion of a Stainless Steel with 13% Cr and 8% Ni in Acids and Mixtures of Acids. W. Machu and M. G. Fouad. (*Arch. Eisenhüttenwesen*, 1957, **28**, Mar., 157–165). The weight losses of the steel in HCl, H_2SO_4, and H_3PO_4, and in mixtures of these acids, and in the presence and absence of organic and inorganic inhibitors have been studied. The determining factor in corrosion and inhibiting effect was found to be the ability to form or destroy the inhibiting layer formed by adsorption or chemisorption.

Some Aspects of the Stress Corrosion of Steel in Caustic Soda Solutions. F. A. Champion. (*Chem. Indust.*, 1957, July 13, 967–975). Plant extracting Al by the Bayer process is dealt with.

The Formation of Scale. P. Gröbner. (*Hutník*, 1956, **6**, (12), 364–366). [In Czech]. The structure and kinetics of formation of scale on iron and steel are discussed.—P. F.

Oxidation of Iron–Nickel Alloys. II. Electron Diffraction at High Temperature. R. T. Foley, C. J. Guare, and H. R. Schmidt. (*J. Electrochem. Soc.*, 1957, **104**, July, 413–417). Alloys with 30, 41, and 78% Ni were oxidized at 600–900° and examined by electron diffraction in a special furnace. The predominant oxide on 30 and 41% alloys was a ferrite, at 700 and 900° NiO and MoO_3 exist on the 78% alloy (with 3·8% Mo). Lattice spacings of the alloy structure were sometimes carried over into the first oxide film formed, later growth gives values for the oxide corresponding to X-ray findings.

Researches on Corrosion and Inhibition. Adsorption, Inhibition and the Langmuir Equation. G. S. Gardner. (*J. Franklin Inst.*, 1957, **263**, June, 523–535). Corrosion of steel has been studied in dilute acetic acid in presence of an oil phase and with and without organic inhibitors. The Langmuir constant b is suggested as an index of inhibitor behaviour.

Corrosion of Stainless Steels in Supercritical Water. W. K. Boyd and H. A. Pray. (*Corrosion*, 1957, **13**, June, 375–384). Hardenable and nonhardenable grades were tested in degassed supercritical water at 800, 1000, and 1350° F at 5000 psi. Intergranular corrosion occurred at 1350° while at 1000° F only Inconel X showed penetration along grain boundaries. Phase changes occurred only at 1350° F. All had good resistance at 800° F. AISI Type 316 material showed stress-corrosion cracking at 1350° F.

Dissolution of Iron in Acidified Chromic Chloride Solutions. R. Skomoroski and C. V. King. (*J. Electrochem. Soc.*, 1957, **104**, July, 417–419). The rate of dissolution from a rotating

cylinder was studied in 4M HCl with purple and green $CrCl_3$ and $AlCl_3$ and NaCl up to 0·2M. Chromic ion is not reduced. Cathodic polarization did not reduce chromic ion but protected the iron. Iron catalyses oxidation of chromous ion by H+.

The Activation and Passivation of Stainless Steel in Aerated Sulphuric Acid. G. H. Cartledge. (*J. Electrochem. Soc.*, 1957, **104**, July, 420–426). Potential-time curves for Nb-stabilized steel at 85° C during its transition to the state stable to aerated 0·1N H_2SO_4. Characteristic inflexions were found and provisionally related to the processes occurring. A critical recovery potential was found above which spontaneous passivation occurs, and approximately equal to the hydrogen potential.

Electrochemical Properties of Alloys. VII. On the Anodic Behaviour of Iron–Chromium–Nickel Ternary Alloys in Sulphuric Acid Solution. S. Morioka and K. Sakiyama. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, July, 438–442). [In Japanese]. Results are given for alloys containing 12–23% Cr and 0–12% Ni in 5% acid solution. The higher the Ni content, the smaller the critical c.d. for passivity. Alloys of uniform γ-phase were most favourable for passivation. The strong tendency of the high-Cr alloys to remain in the passive state is reduced by a small amount of Ni.—K. E. J.

Some Case Histories of Stress Corrosion Cracking of Austenitic Stainless Steels Associated with Chlorides. H. R. Copson and C. F. Cheng. (*Corrosion*, 1957, **13**, June, 397–404). Cases in water, steam, brines and other solutions are quoted, always with high chloride concentration either initially or by accumulation. The conditions were usually hot, acid, and with air present. Characteristic transcrystalline cracks were always formed and were due to internal tensile stress in presence of chlorides. Nickel content increases the resistance and Inconel is practically immune to stress corrosion cracking.

Periodic Fluctuations in Current Strength in the Anodic Solution of Zinc. K. Schwabe. (*Metalloberflache*, 1957, **11**, Jan., 1–6; Feb., 37–40). The effects of initial current density and temperature on the duration, frequency and amplitude of periodic fluctuations in current strength during the anodic solution of zinc in various aqueous solutions were discussed, and it is shown that the fluctuations are due to the formation of a highly viscous layer of solid salts.—L. D. H.

Evaluation of Surface Treatments for Low-Alloy Steels. Part I: Test Method for Heat-Resistant Corrosion Protective Coatings on Steel. Part II: Paint Chromizing, Paint Siliconizing and Coating of Titanium–Boron Low-Alloy Steels. S. Tour. (*PB*.121087, 121088, 1954, pp. 28, 19, *TIDU List*, 751). Equipment for testing at 800° F in a corrosive atmosphere is described, and a condensation test of cooling and warming in dry and moist air is given with tests on 16 coatings. The Cr and Si diffusion process by heat-treatment after painting with Cr or Si powders or their Fe alloys. The Si coating is hard and brittle but the chromized parts are ductile and weldable.

Longer Life for Steel Products. BISRA's Work on Corrosion and Coating Control. (*Brit. Steelmaker*, 1957, **23**, July, 216–218). An account based generally on exhibits at the laboratories' exhibitions of work.

Choice of Corrosion Resisting Materials in Extraction of Metals. K. P. Mukherjee and A. K. Lahiri. (*Symposium on Mineral Beneficiation and Extractive Metallurgical Techniques, Council Sci. Ind. Res. Jemshedpur, India*, 1957, Feb. 8–11, pp. 22). In roasting, leaching, etc., high-Cr or Cr–Ni steels are usually used. These and other steels are reviewed with regard to the various uses, and non-ferrous and non-metallic materials used for acids, alkalis and other corrosive liquids are also mentioned.

Corrosion Resistance of Chromized and Unchromized Alloys in Contact with Burning High Sulphur Coal. (*Canad. Mines Tech. Surveys, Research and Special Projects*, (July-Dec. 1956), 1957, Feb. 8, 23–25). Stainless steels are included.

Rust Preventive Properties of Aeronautical Greases. S. M. Collegeman. (*PB* 119122, (no date), *TIDU List* 739, pp. 12). Roller bearings were stored at high humidity or sprayed with water for 1 h while rotating and stored at room temperature. 2 of 13 greases failed these tests.

Shipboard Evaluation of Zinc Galvanic Anodes Showing the Effect of Iron, Aluminium and Cadmium on Anode Performance. E. C. Reichard and T. J. Lennox jun. (*Corrosion*, 1957, **13**, June, 410–416). Addition of Al + Cd improves performance and increases tolerance for iron in the anodes. Tests on ships, either active or in harbour, are quoted.

Corrosion Protection of Steel. R. A. Schamschula. (*Maschinenwelt Elektrotechnik*, 1957, **12**, Jan., 5–9). A general survey is given of methods of protecting steel against corrosion. (30 ref.).—L. D. H.

Anodic Corrosion Protection of Steel. M. Pražák. (*Hutnické Listy*, 1956, **11**, (11), 644–648). [In Czech]. A study was made of the anodic protection of 18/8 stainless steel in strong acids by means of cathodes based on Fe_2O_3, FeO, and MnO_2. The conditions under which these cathodes maintain the steel in the passivated state are given. The protection afforded by these semi-conducting materials was found to be very good, and the method is considered to be most promising and economical for use in practice.—P. F.

An Automatic Regulator for Cathodic Protection of Pipe Lines. V. I. Ivanenko and A. D. Pushchalovskii. (*Avtomatika*, 1957, (1), 18–26). Two designs are given for protective systems for gas pipes. A thermogenerator is recommended as a d.c. source.

ANALYSIS

A New Process for the Sampling of Killed and Unkilled Liquid Steel for the Purpose of Oxygen and Hydrogen Determination. H. Schenck, K. H. Gerdom, and K.-G. Schmitz. (*Arch. Eisenhüttenwesen*, 1957, **28**, Mar., 123–125). Quartz pipettes, evacuated to $1 \cdot 10^{-4}$ mm and sealed with soft iron tops, are used to obtain the total gas content of the molten steel. A rapid oxygen determination can be carried out with un-killed samples.—L. J. L.

Determination of the Carbon Content of Steel from the Hardness of a Hardened Sample. M. G. Pyatigorskiĭ. (*Zavodskaya Laboratoriya*, 1956, **22**, (7), 778–781). [In Russian]. In the method described the sample of liquid carbon or alloy structural steel is poured into a special steel mould in which its rate of cooling is retarded by the presence of an air annulus. The sample (at 1100° C approximately) is immediately removed from the mould and quenched in iced water. Two of the parallel faces of the quenched sample are then subjected to Rockwell hardness tests, from the results of which the carbon content (range up to 0·52%) is found. A determination takes 1·5–2 min.—S. K.

On Coulometric Titration. Part 6. Coulometric Titration of Iron (II) with Electrolytically Generated Manganese (III). F. Kawamura and S. Suzuki. (*Bull. Fac. Eng. Yokohama Nat. Univ.*, 1957, **6**, Mar., 95–98).

A New Analytical Method for the Determination of the Iron in Chromium Baths. H. W. Dettner. (*Metalloberfläche*, 1957, **11**, Jan., 12). Iron is determined without separation from the chromium bath by adjusting the pH value to 2·5, and titrating with E.D.T.A. solution, using disodium pyrocatechol-3, 5-disulphonate as indicator. The maximum iron content is 20 g/litre.—L. D. H.

Determination of Metals by Differential Polarimetry. M. Freegarde. (*Chem. Indust.*, 1957, July 13, 984–985). Changes in the rotation of tartrate solutions in presence of iron (and silver) can be used.

The Colorimetric Determination of Phosphorus in Steel and Copper-Base Alloys. W. T. Ewell and H. N. Wilson. (*Analyst*, 1957, **82**, June, 453–454). A correction on the interference by As is noted. A method of preliminary separation of As is given.

The Rapid Photometric Determination of Chromium in Steel in the Range between 12% and 20%. J. Scheunemann and B. Strieter. (*Giessereitechn.*, 1957, **3**, Mar., 68–69). A method of determining Cr up to 20% by the Pulfrich photometer is described.—L. J. L.

Colorimetric Determination of Molybdenum in Scheelite Ores and Concentrates. R. P. Hope. (*Anal. Chem.*, 1957, **29**, July, 1053–1055).

Annotated Bibliography of the Analytical Chemistry of Niobium and Tantalum, Jan. 1935–June 1953. F. Cuttitta. (*U.S. Geol. Survey Bull.*, 1957, 1029-A, 1–73).

An Investigation of a Method for the Combined Determination of Niobium and Tantalum. B.I.S.R.A. Methods of Analysis Committee. (*J. Iron Steel Inst.*, 1957, **187**, Dec., 341–343). [This issue].

Polarographic Determination of Niobium and Tungsten In Alloys. Z. S. Mukhina and A. A. Tikhonova. (*Zavodskaya Laboratoriya*, 1956, **22**, (10), 1154–1156). [In Russian]. An account is given of an investigation to find optimal conditions

for the polarographic determination of Nb in Fe-, Ni-, Cr-, and Co-base alloys in the presence of W. The method developed enables both niobium and tungsten to be determined, a 10-N hydrochloric acid solution being used. Either potassium citrate or Trilone are present, and the reduction potentials for Nb and W are 0·28 and 0·42–0·44 V., respectively. Mineral niobium content is 0·25% and a five-fold excess of tungsten is permissible.—S. K.

The Photometric Determination of Arsenic in Commercial Iron by Means of Silver Diethyldithiocarbamate. Z. Vecera and B. Bieber. (*Giessereitechn.*, 1957, **3**, Mar., 61–64). The sample is dissolved in a mixture of HNO_3 and H_2O_2. The arsenic content is determined photometrically by passing the AsH_3 through a pyridine-silver-diethyldithiocarbamate solution, which assumes a vivid red colour. Accuracy is ±0·0004% As for 0·1 g samples with As contents of fractions of 1%.

The Applications of Physical Methods of Analysis in the Gas Industry. A. B. Densham and G. Gough. (*Gas World*, 1957, **146**, July 20, 118–121). Spectoscopic, chromatographic, radiographic, polarographic, and X-ray methods are briefly outlined. (96 references).

Rapid Determination of Carbon, Phosphorus and Sulphur. (*Metallurgia*, 1957, **55**, May, 256). A short description is given of results obtained with the new Fluorite Polychromator built by Hilger and Watts Ltd. The accuracy achieved is excellent and standard deviations of analyses for C, S, and P at different levels are shown.—B. G. B.

Spectrographic Analysis of Low-Alloy Steel. A Plate Calibration Technique Based on Relative Intensities of Standard Iron Lines. T. G. Shamanna and N. P. Gandhi. (*Trans. Indian Inst. Met.*, 1954–55, **8**, 309–317). Tests on a wide range of low alloy steels have confirmed that the B.I.S.R.A. recommended procedure of using a group of iron lines as internal standards gives accurate results.—A. D. H.

Spectrographic Analysis of High-Alloy Steels. T. G. Shamanna and P. N. Gandhi. (*Trans. Indian Inst. Met.*, 1954–55, **8**, 319–329; discussion 329–332). The concentration-ratio method of spectrographic analysis was applied to high-speed, stainless, and low alloy steels. It was concluded that the method is accurate for all these materials.—A. D. H.

A Photoelectric Spectrometer for the Study of Steels. H. Diebel and W. Hanle. (*Arch. Eisenhuttenwesen*, 1957, **28**, Mar., 127–132). The construction and operation of the instrument are described. The excitation of the steel samples is effected by a multisource-type spark discharge. The instrument is used for the rapid determination of Si, Mn, Cr, Ni, Mo, and Cu in a steel works. It can also be used for the analysis of light-metal alloys.—L. J. L.

Practical Experience with the New Photoelectric Lattice Spectrometer. F. Thorn. (*Arch. Eisenhüttenwesen*, 1957, **28**, Mar., 133–134). Operating experience with a spectrometer used since January 1956 for the determination of Si, Mn, Cr, Ni, Mo, and Cu in unalloyed and low-alloy steels is reported.

The Further Development of Photoelectric Emission Spectrum Analysis. H. Krempl and G. Scheibe. (*Arch. Eisenhüttenwesen*, 1957, **28**, Mar., 135–139). Processes occurring at the spectrochemical light source can be studied by observing spark curves in different gases in conjunction with the detrition of material and by analysing light emission in individual sparks. In sparking, there is a close connection between the discharge current and the sequence in time of ionic and atomic lines. The requirements a versatile instrument should satisfy are discussed, and an existing instrument is described that has been successfully used in the analysis of metalloids in steel.—L. J. L.

Electron Probe Microanalyzer and Its Application to Ferrous Metallurgy. R. Castaing, J. Philibert, and C. Crussard. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, Apr., Section 2, 389–394). A description is given of the principle, design, and operation of the electron probe microanalyser, which consists of three main parts: an electron optical system, an X-ray spectrograph, and an optical microscope. The first results obtained are surveyed, demonstrating its variety of applications. These include local analysis of non-conductors, metals, and alloys; measurement of composition variation within one phase or segregate; determination of the composition of non-metallic inclusions; and estimation of distribution coefficients of various elements between precipitates and matrix.—G. F.

Method of Contact Micro X-Radiography of Alloy Steels and Alloys. B. A. Movchan. (*Zavodskaya Laboratoriya*, 1956, **22**, (7), 817–820). [In Russian]. The physical bases of the micro X-radiographical method are briefly discussed and improvements by which its scope has been enlarged are described. To reduce to a minimum the deleterious effect of the continuous spectrum special X-ray tubes and optimal voltages were used. The selection of optimal section thickness is discussed and data on atomic absorption coefficients for various anode materials are presented. A microradiograph of an Fe–Ni (8%)–Nb (1%) at × 120 magnification is given as an example of the suitability of the method for alloy steel.—S. K.

The Application of Radioactive Isotopes and Radiation Sources to Metallurgy. A. M. Samarin. (*Neue Hütte*, 1957, **2**, Feb.-Mar., 69–76). A discussion of the use, in the U.S.S.R., of radioactive isotopes in studying (1) metallurgical reactions, (2) blast furnace operations, (3) casting processes, (4) diffusion phenomena, and (5) the control of metallurgical processes.

INDUSTRIAL USES AND APPLICATIONS

Fundamental Aspects of the Economic Utilisation of Alloy Steels in Screw Production. F. Beneš. (*Hutnické Listy*, 1956, **11**, (11), 641–643). [In Czech]. The suitability of several typical structural steels for screw production by cold pressing was experimentally investigated with a view to the possibility of reducing the number of types of steels used for that purpose to a minimum. Hardenability and mechanical properties of the steels are given, and it is shown that a considerable advance towards the aims of the work was possible on the basis of the results obtained.—P. F.

Steel as a Bearing Material. O. Gersdorfer. (*Maschinenwelt Elektrotechnik*, 1956, **11**, Sept., 279–281). The properties and applications of steel bearing surfaces are discussed.—L. D. H.

Role of Stainless Tubing in Atomics. N. D. Groves. (*Steel*, 1957, **140**, May 20, 143–144). Quality demands for stainless steel pipe and tubing for atomic energy applications are severe; comment is made on methods of ensuring suitable products.—D. L. C. P.

New Fastener Stays Strong at High Temperatures. (*Iron Age*, 1957, **179**, May 16, 110). Properties of the HiTm9 fastener are described. It is a bolt for application up to 900° F in high speed aircraft, but other applications are envisaged. The material is Vascojet 1000, with a diffused Ni–Cd plating. Tensile strength at 900° F is 170,000 p.s.i.—D. L. C. P.

HISTORICAL

Blast Furnace USA: 1. Colonial Ironmakers. M. O. Holowaty and C. M. Squarcy. (*J. Met.*, 1957, **9**, May, 656–661). The development of the blast furnace and early iron production in U.S.A. are reviewed. Particular mention is made of the Spotswood furnace at Germanna (1716) and the Principio Co. in Maryland (1714).—G. F.

Metallurgical Problems Encountered in the Construction of a Steel Bridge 130 Years Ago. H. Diekmann. (*Stahl u. Eisen*, 1957, **77**, May 2, 581–586). An historical account of the testing of the structural parts of a suspension bridge over the river Lahn at Nassau in Germany, built in 1830.—T. G.

ECONOMICS AND STATISTICS

The American Steel Industry. 1. The Economic Background. K. Kyle. **2. The Outlook for Labour.** T. Campbell. **3. Management and Organization.** N. R. Wills. **4. Technological Trends.** C. M. Parker. **5. Looking into the Future.** B. F. Fairless. (*Steel Rev.*, 1957, July, 24–52). A review.

The Influence of Modern Technological Processes on Production Costs in the Metal Industry. J. Hlaváček. (*Hutník*, 1956, **6**, (12), 353–357). [In Czech]. Increases in efficiency and output rates attainable by the use of modern agglomeration processes for blast-furnace ores, by the application of O_2 in converter practice, by increases in the capacities of blast furnaces, and by the introduction of continuous casting of steel are discussed.—P. F.

Building and the Steel Industry. I. Bowen. (*Steel Rev.*, 1957, July, 19–23). A statistical estimate of requirements.

Rail Transport Cost Accounting for Metallurgical Works. A. M. Genesin, B. A. Dorfman, and P. M. Poplavskii. (*Stal'*, 1957, (1), 76–79). [In Russian]. Rail transport costs in iron and steel works are analysed and compared for several Soviet

works. It is suggested that for proper comparison to be made the t-km must be replaced by a cost per t of freight transported.—S. K.

Transition to a Seven-Hour Day for Workers and Staff. L. A. Bergauz. (*Stal'*, 1956, (11), 1026–1031). [In Russian]. In the course of the present five-year plan the working day of iron and steel and other workers and staff under the Ministry of Ferrous Metallurgy is to be reduced gradually to seven or six hours. This transition is considered in detail in the present article. Further contributions are invited in an editorial note.—S. K.

Some Problems in the Methodology of Calculating Labour Productivity. (*Stal'*, 1956, (11), 1032–1034). [In Russian]. Points raised in a recent article by Ya. D. Kats on labour-productivity calculation are critically considered by G. P. Zhuravlev (1032), I. F. Khoroshilov (1032–1033) and K. A. Popov (1033–1034). The main objection is to Kats' view on what production should be included in the productivity calculation.—S. K.

MISCELLANEOUS

Observations on the Magnetic Transition in Hematite at − 15° C. G. Haigh. (*Phil. Mag.*, 1957, **2**, 8th Series, July, 877–890).

The Recovery of Manganous Sulphate from Leach Liquors. (*Canad. Mines Tech. Surveys, Research and Special Projects (July-Dec.* 1956), 1957, Feb. 8, 4–5).

BOOK NOTICES

GREEN, H. L., and W. R. LANE. "*Particulate Clouds: Dusts, Smokes and Mists.*" Their physics and physical chemistry and industrial and environmental aspects. La. 8vo, pp. xix + 425. Illustrated. London, 1957: E. & F. N. Spon, Ltd. (Price 70*s.*).

This is a valuable book for all concerned with the study of various aspects of the properties and behaviour of particles or suspensions of particles. It is divided into two parts. Part I deals with the theoretical aspects of individual particles and clouds of suspended particles and their application in research. In view of the tremendous progress in the study of particles in recent years there was an urgent need to produce a comprehensive and systematic digest of the scattered publications. This object has been successfully achieved. The standard of the treatment is high, and this part of the book will be of particular value to research workers. Part II, dealing with industrial and environmental aspects, is of wider general interest and includes some basic information about industrial gas cleaners, atmospheric pollution and practical uses of particulate clouds.

Part I covers a wide variety of subjects ranging from the mechanisms of formation of particulate clouds, equipment used for generating them for research purposes, physical characteristics of particles (such as rate of fall and optical properties) to the fundamental processes of the removal of particles from gases and methods of counting, sizing and sampling. It is unfortunate that in the section dealing with cyclone separators (Chapter 6) C. J. Stairmand's useful work on the subject ("The Design and Performance of Cyclone Separators," *Trans. Inst. Chem. Eng.*, 1951, **29**, 356) is not mentioned. Few errors have been found, one of them being in Table 7.3 (p. 247); the figures in the second column should be 2 and $\frac{2}{3}$ instead of 1·5 and 0·5 respectively.

The topics considered in Part II range from gas cleaning and sampling on industrial plants, health hazards, some aspects of atmospheric pollution (including smog) to the natural aerosols (clouds, fog) and the use of particulate clouds in industry, agriculture, medicine and war.

The authors have succeeded in giving an up-to-date account of a vast subject. Extensive lists of references will enable readers to obtain further information on topics of specific interest.—S. BADZIOCH.

DAVIES, OWEN L. (Editor). "*Statistical Methods in Research and Production.*" With Special Reference to the Chemical Industry. Third edition (revised and enlarged). Authors: George E. P. Box, Wilfred R. Cousins, Owen L. Davies, Francis R. Himsworth, Harold Kenney, Maurice Milbourn, William Spendley, Wilfred L. Stevens. La. 8vo, pp. x + 396. Illustrated. London, Edinburgh, 1957: Oliver and Boyd (for Imperial Chemical Industries, Ltd.). (Price 45*s.*).

The latest edition of this well-known statistical handbook is now slightly larger, and considerable revision has been carried out; no doubt one reason for this is the addition of Dr. Box to the team of authors. The chapters on elementary techniques are substantially the same as in the earlier editions, as are the chapters dealing with contingency tables and control charts. However, in discussing significance tests emphasis is now placed on the calculation of confidence intervals, and in consequence the chapters concerned have been completely rewritten. The same idea is developed in the revised chapters on the analysis of variance. The technique of regression is now considered more fully and, in the reviewer's opinion, with greater clarity. A new chapter deals with the cost of experimentation, and is well illustrated with examples. The problem of reaching a balance between precision and expense is vital to industrial experimentation, and this chapter may well be the most important single addition to the book. Finally, a description of sequential sampling procedures has been added to the chapter on inspection sampling and quality control.

The handbook is still very easy to understand and should be of greatest use to those people who wish to make statistics a tool rather than a study. But the general principles necessary to the full interpretation of any data are a little understated in the book. In this respect only the book should not be regarded as self-sufficient.—P. THURSFIELD.

LANG, RUTH (Editor). "*Laboratory and Workshop Notes, 1953–1955.*" A fourth selection reprinted from the *Journal of Scientific Instruments*, compiled for the Institute of Physics. 8vo, pp. xii + 248. Illustrated. London, 1957: Edward Arnold (Publishers) Ltd. (Price 30*s.*).

Dr. Lang is to be congratulated on compiling another collection of a hundred and twenty notes, each one of which has a high potential value to the experimental scientist in almost any field of physical science or engineering. In these days of expanding technology and narrower individual specialization, it becomes increasingly difficult to ensure that the maximum of research or development can be carried out with the minimum of misapplied effort. This book, with earlier volumes in the same series, presents in a compact form a number of what one might almost term "labour-saving gadgets" for the laboratory. Thus one can readily acquire practical details of what are often unfamiliar techniques, details which are all too commonly omitted from standard reference works.

While I found the table of contents and indexing fully adequate for this particular volume, it would seem that the value of this book, in relation to earlier volumes, would be enhanced by a comprehensive index covering the whole series. However, this is perhaps an unfair criticism of a book which, in its very title, has a specific remit to cover a limited field which it does most effectively.

Finally, it must be pointed out that, whilst the direct value of each of these notes may be high, a most useful additional contribution comes from the way in which they may stimulate an unrelated line of thought into the solution of a particular problem. This stimulus may simply take the form of suggesting a familiar tool or material for a hitherto unfamiliar purpose. Thus I regard a book of this type as an excellent breeding-ground for ideas.

VEREIN DEUTSCHER EISENHUTTENLEUTE. "*Stahldrahterzeugnisse.*" Herausgegeben vom Ausschuss für Drahtverarbeitung im Verein Deutscher Eisenhüttenleute. 8vo. Illustrated. Band I, pp. xii + 340. Band II, pp. x + 315. Düsseldorf, 1956: Verlag Stahleisen m.b.H. (Price DM.70.– the two volumes).

In 1941 the Verein Deutscher Eisenhüttenleute, in collaboration with the German Steel Wire Manufacturers Association, decided to produce a text book on the manufacture of steel wire products. The planning of the work was undertaken by the late Professor Anton Pomp and continued, after his death in 1953, by Dipl. Ing. Otto Peltzer and Dr. Ing. Werner Lueg. As would be expected

from the calibre of these authorities, the work now published in two volumes is quite outstanding and probably unique in the literature of the wire industry.

No attempt is made to describe wiredrawing processes except where necessary in connection with some particular manufactured product. The two volumes do, however, cover in great detail the production and properties of the principal products manufactured from wire, and are divided into sections written by acknowledged authorities in each particular field. The treatment of each product follows a similar and logical pattern: (1) Historical development; (2) Raw materials; (3) Characteristics and properties of the product; (4) Industrial application; (5) Machinery; (6) Methods of testing, standards, bibliography.

Volume I contains sections on wire ropes, welding wire, welded wire mesh, steel wire for concrete reinforcement, wire working and allied products, tyre beads, steel wool, and steel shot to name a few. The sections on wire ropes and welding electrodes are particularly comprehensive. Volume II includes cold formed helical springs, upholstery springs, needles, chains, rivets, screws, wire nails, barbed wire, wire netting and similar agricultural products; saw wire, woven wire cloth, products for textile machinery, card wire.

Both volumes are very detailed and up-to-date and are magnificently illustrated by photographs, line drawings and sketches. The Verein Deutscher Eisenhüttenleute is to be complimented on the production of such a comprehensive work; it will be of great value to all interested in the fabrication and use of steel wire in its various forms.

WILTSHIRE, W. J. (Editor). "*A Further Handbook of Industrial Radiography.*" By Members of the Non-Destructive Testing Group of the Institute of Physics. 8vo, pp. viii + 334. Illustrated. London, 1957: Edward Arnold (Publishers), Ltd. (Price 50*s.*).

This book is intended as a complement to the "Handbook of Industrial Radiology" published in 1944 and consists of a series of independent review articles primarily on various fundamental aspects and new developments which have been advanced in recent years. The quality and scope of these reviews is somewhat variable and the authors appear to differ widely in their conception of the purpose of the book. Despite this unevenness and the inevitable lack of orderly development in a book of this type, it provides useful reviews of several important aspects of radiological examination and the articles are generally copiously referenced.

The fundamental physical processes involved in radiography are rather fully discussed in articles entitled "Scattered Radiation" and "Radiographic Quality and Sensitivity," while on the more technical aspects there is an extremely useful and concise review entitled "Natural and Artificial Gamma Ray Sources" and others on fluoroscopy, film processing, weld radiography, radiography of castings, etc. On the more recent developments in techniques and equipment there is a review of megavolt X-ray generators with a somewhat inadequate appraisal of their radiographic potentialities, and an interesting article on xeroradiography, together with a general introduction to electrical detection and measurement techniques.

Ancillary subjects which have been included cover flash radiography, microradiography, and applied X-ray diffraction methods. Two of these ancillary reviews together with an article on stereoradiography could have been curtailed without serious restriction of their value.

In a brief review of this type it is not possible to discuss the technical data in detail, but mention must be made of the flaw sensitivities quoted in the article on fluoroscopy. Even allowing for the author's severe criteria of flaw visibility, sensitivity figures of the order of 22 to 40% for steel are discouragingly poor and would indicate that such techniques are of little practical value. This, we know, is not so in practice, and the sensitivities attainable bear far more favourable comparison with the values obtained with image intensifiers than the author suggests, particularly if projective magnification is used.

This book is written primarily from the physicist's point of view, but this does not imply that technologists, particularly those who are interested in the physical aspects of these techniques, will not find the book a valuable source of information.—B. D. BURNS.

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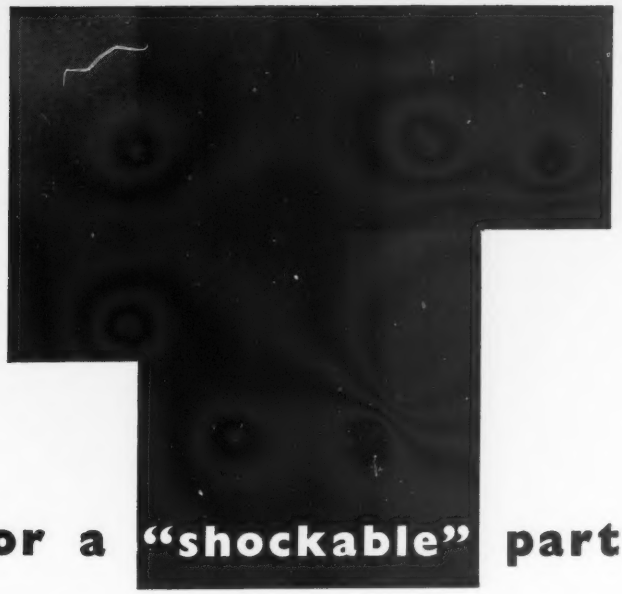
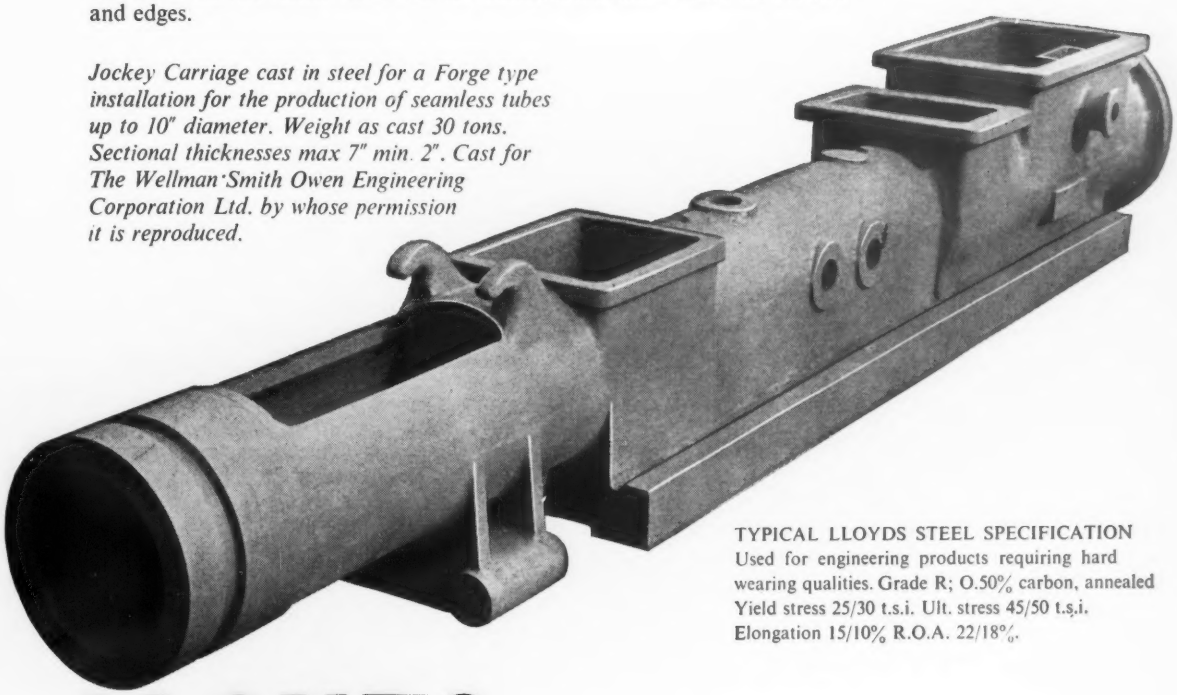


cast for a "shockable" part

This jockey carriage is cast in steel for a forge type of installation for the production of seamless tubes up to 10" diameter, and has to withstand unpredictable shock load. It is a good example of the versatility of steel casting, in that its sectional thickness varies from 2" to 7", essential requirements that make it almost impossible to fabricate by any other method.
Owing to good method planning this casting is free from defects and of especial interest is the fine finish. Note also the sectional thickness round the corners and edges.

Jockey Carriage cast in steel for a Forge type installation for the production of seamless tubes up to 10" diameter. Weight as cast 30 tons. Sectional thicknesses max 7" min. 2". Cast for The Wellman·Smith Owen Engineering Corporation Ltd. by whose permission it is reproduced.



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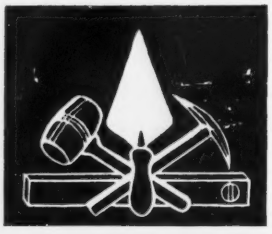

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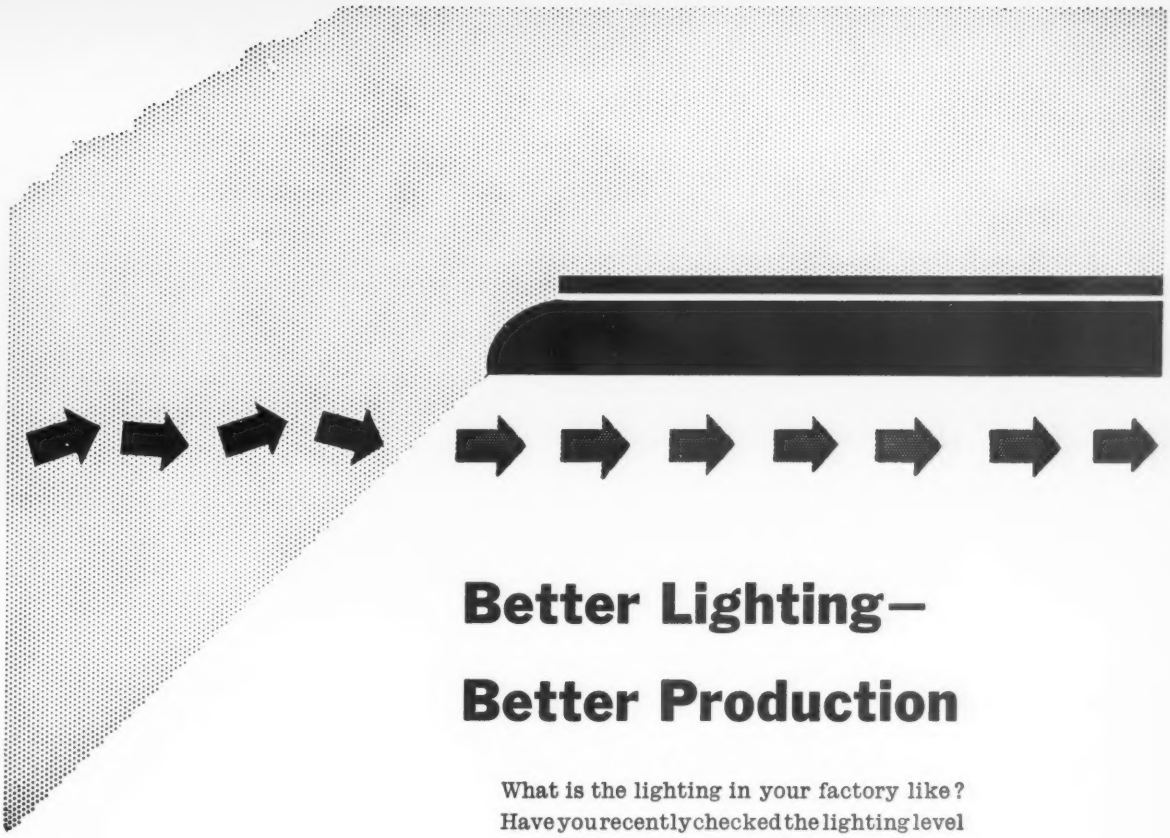

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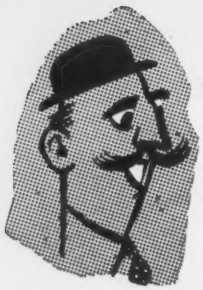

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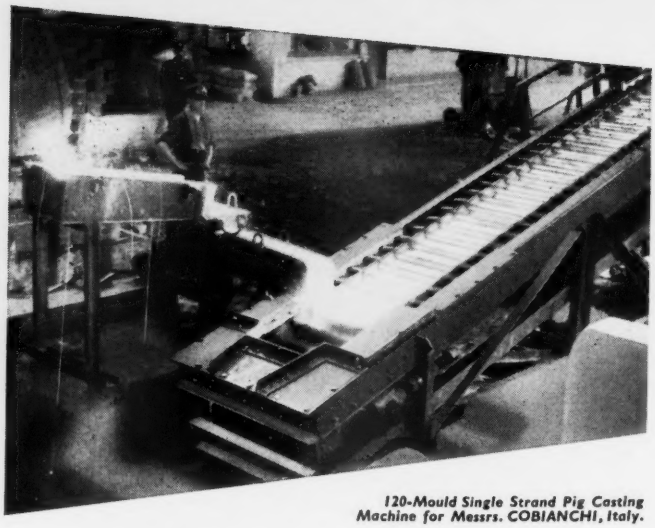

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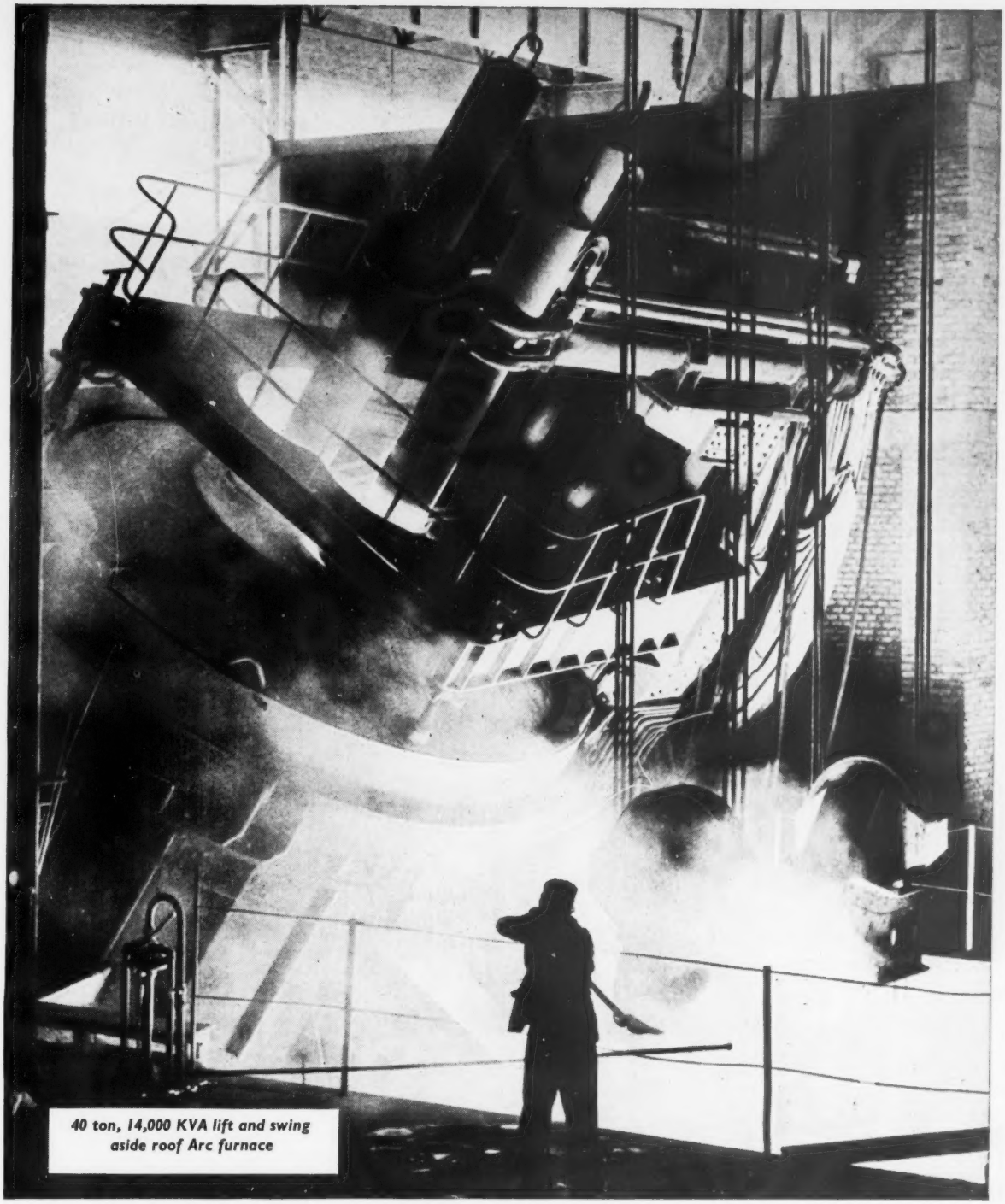

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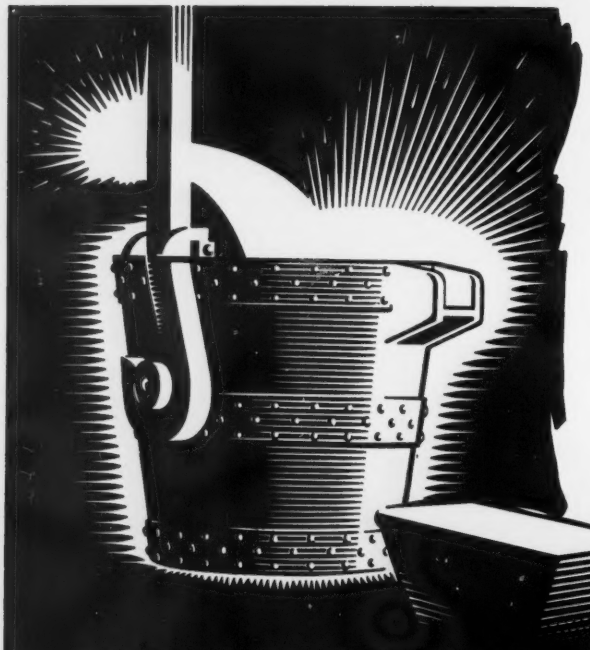

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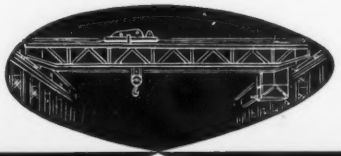

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THE IRON AND STEEL INSTITUTE

Special Report No. 59

COMBUSTION AND HEAT TRANSFER IN AN OPEN-HEARTH FURNACE

A RECENT addition to the Special Report series of the Institute is entitled "Combustion and Heat Transfer in an Open-Hearth Furnace," by S. W. Pearson, M. W. Thring, and J. H. Chesters. This Report describes a large-scale trial carried out on a Templeborough furnace with a team consisting of groups from Messrs. Steel, Peech and Tozer, the Research and Development Department of the United Steel Companies Ltd., the University of Sheffield, and the British Iron and Steel Research Association.

The previous Templeborough trials, reported in Special Report No. 37 (1946), emphasized the necessity for aerodynamic studies, continuous roof-temperature measurement, and the maintenance of the highest practicable furnace pressure. The shift from producer gas to oil firing in the last ten years has added new objectives, e.g. the determination of optimum nozzle size, steam/oil ratio, and burner direction. The work of the International Flame Radiation Committee at Ijmuiden has given preliminary answers to some of the questions, whilst the theory of heat transfer initially expounded in Special Report No. 37 has been developed to a point where calculated melting times and fuel consumptions are of the same order as those found in practice. The main purpose of the present trial was to measure as many factors as possible, leaving only one free parameter in each verifiable part of the calculation; this could then be determined from the observed result and the value thus obtained used with some confidence in future calculations. A subsidiary objective was the comparison of flow pattern and mixing in model and furnace.

The Report, No. 59 in the series, is fully illustrated with diagrams and photographs and is bound in cloth with stiff board covers. Copies may be obtained on application to the Secretary, The Iron and Steel Institute, 4 Grosvenor Gardens, London, S.W.1, price 37*s.* 6*d.* (Members 25*s.*).

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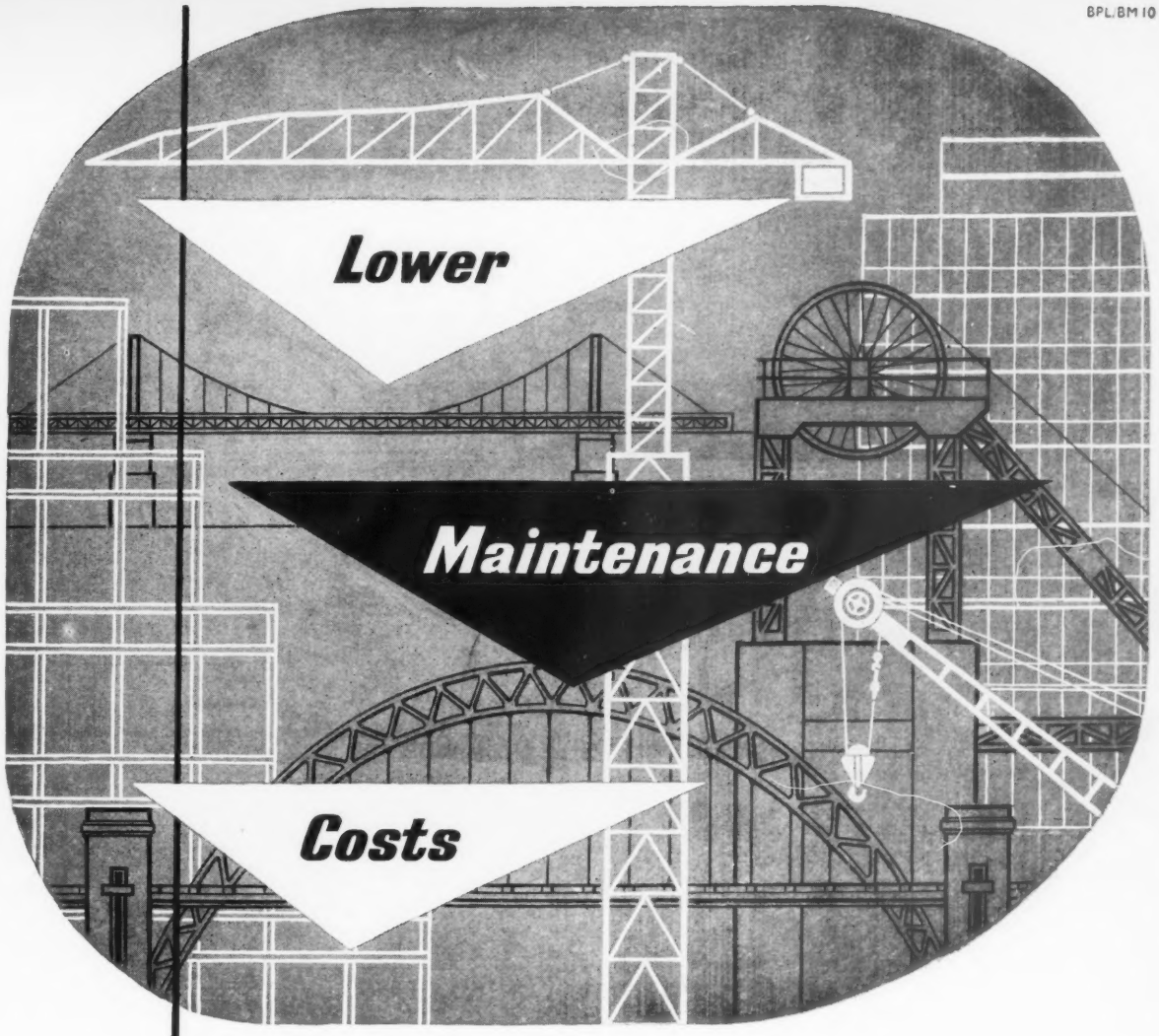

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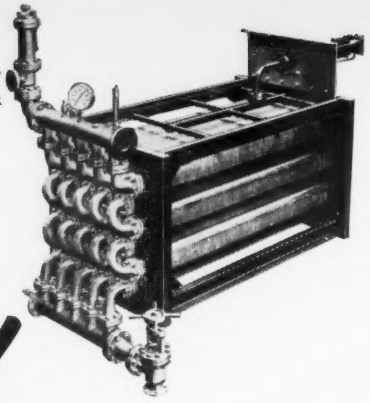

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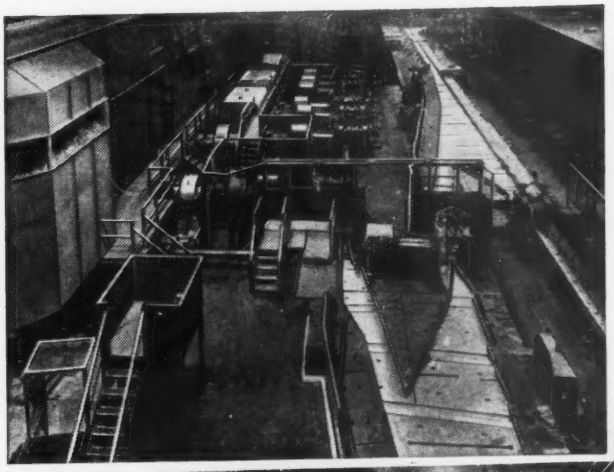

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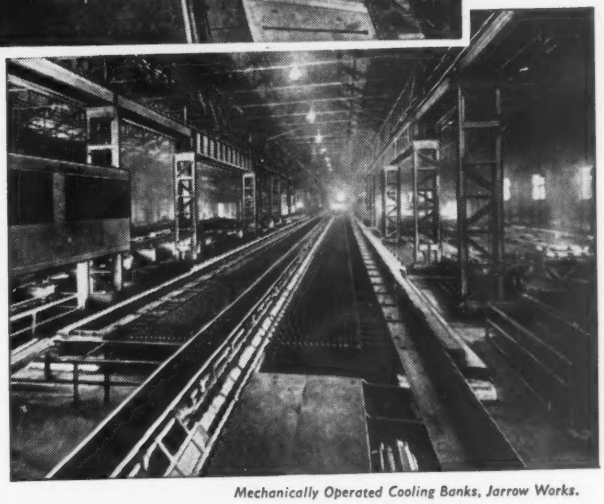

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THE IRON AND STEEL INSTITUTE

Special Report No. 60

Proceedings of the

FORGEMASTERS' MEETING, 1954

In October, 1954, The Iron and Steel Institute held a Joint Meeting with the National Forgemasters' Association and La Chambre Syndicale de la Grosse Forge Française. Technical Sessions were held in London at which nine papers on forging theory and practice and forging plant were presented and discussed. The remainder of the Meeting comprised visits to the English Steel Corporation Ltd., Thos. Firth and John Brown Ltd., and Hadfields Ltd., in Sheffield.

The Proceedings at the Technical Sessions have now been published as a special Report. The five French papers and discussion are presented in English, but summaries of all the papers are also given in French. The Report, No. 60 in the Special Report series of the Institute, can be obtained from the Offices of the Institute, 4 Grosvenor Gardens, London, S.W.1, price 37*s.* 6*d.* (Members 25*s.*).

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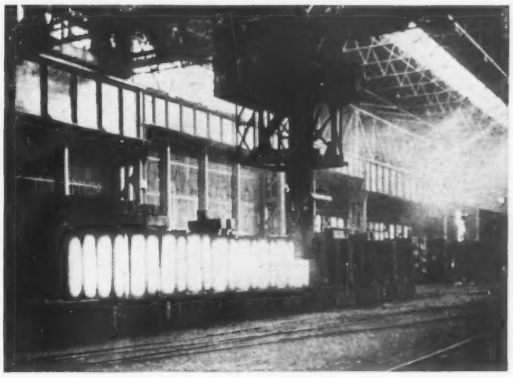

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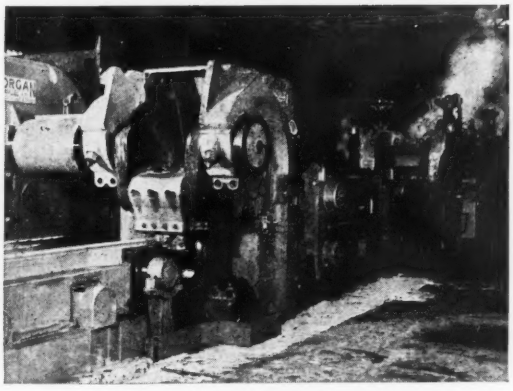

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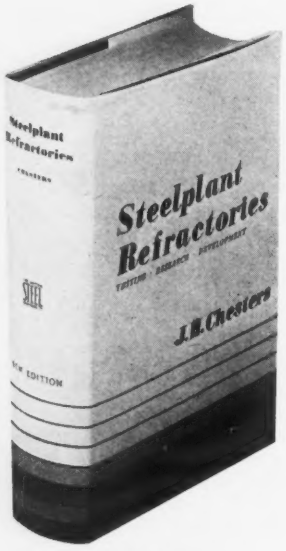

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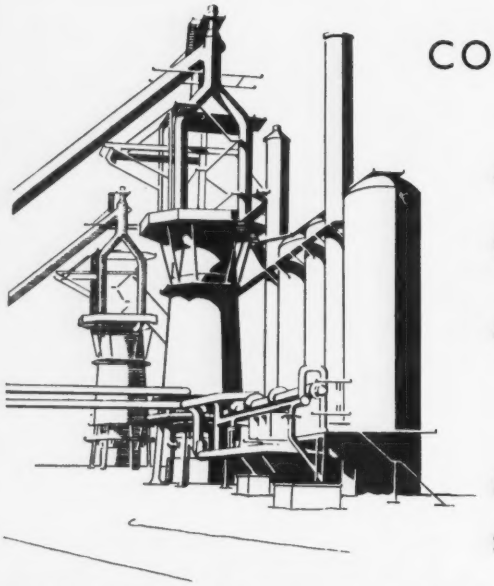

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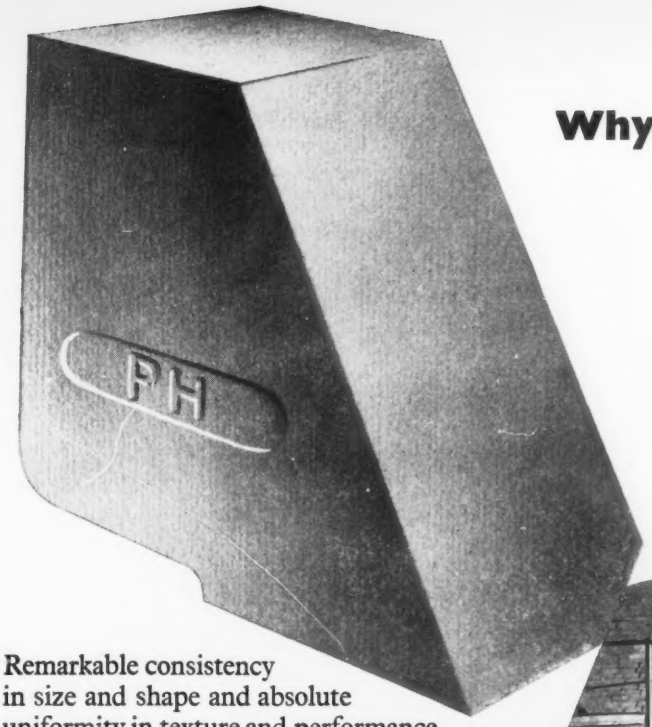

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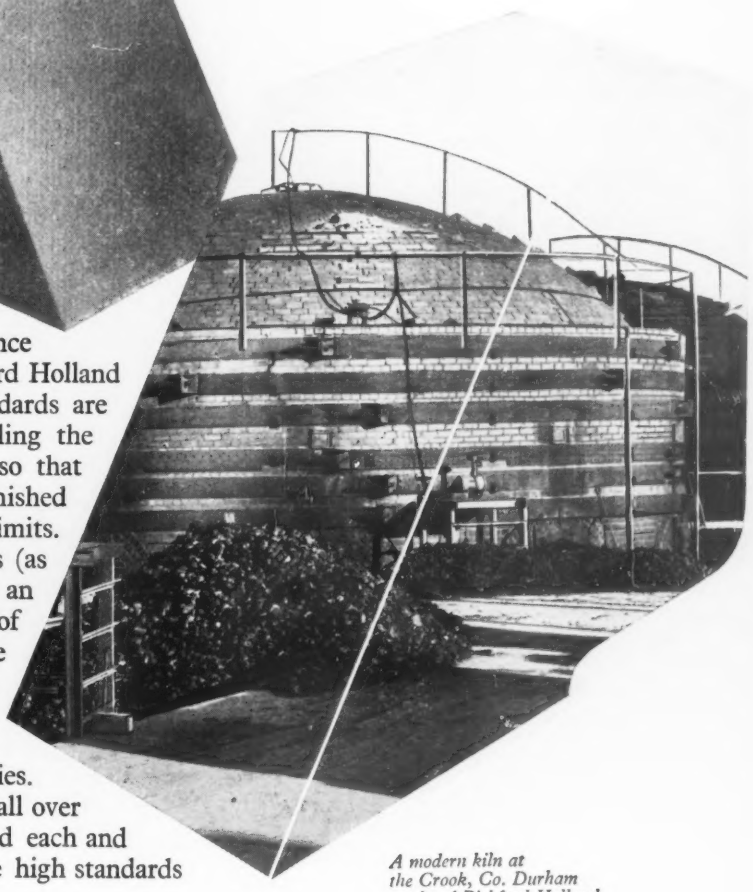

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Index to Advertisers

Acheson Colloids Ltd. ... ... —
Adamson-Alliance Co.. Ltd. 43
Adamson Joseph and Co., Ltd. 69
Allen West & Co., Ltd. 3
Armstrong Whitworth (Metal Industries) Ltd. ... ... —
Arrol, Sir Wm., & Co., Ltd ... ... —
Ashmore, Benson, Pease & Co. Inside Back Cover
Austro—American Magnesite Co. ... 9

Bairds & Scottish Steel, Ltd. 74
Baker Platinum Division ... ... —
Beardmore, William, & Co., Ltd. ... ... —
Benjamin Electric Ltd., The ... ... —
Bigwood, Joshua, & Son, Ltd. ... ... —
Birlec, Ltd. 56
Boot, Henry, & Sons Ltd. 20
Booth, Joseph, & Bros. ... ... —
Bowden-Jackson (Constructions) Ltd. ... 64
Bradley & Foster, Ltd. ... ... —
Brightside Foundry & Engineering Co., Ltd., The 48
Bristol's Instrument Co. Ltd. ... ... —
British Acheson Electrodes Ltd. ... ... —
British Electrical Development Association 65
British Electro Metallurgical Co., Ltd. ... 31
British Furnaces, Ltd. 56
British Oxygen Gases, Ltd. Inside Front Cover
British Paints, Ltd. 71
British Rollmakers Corporation, Ltd. ... 45
British Thomson-Houston Co., Ltd., The 37
British Timken, Ltd.... 21
Brown Bayley Steels, Ltd. 50
Burn Fireclay Co. Ltd., The ... ... —
Bushing Co., Ltd., The 42

Carblox Ltd. ... ... —
Clyde Crane & Booth Ltd. ... ... —
Clyde Crane & Engineering Co. ... ... —
Colvilles, Ltd. 12
Consett Iron Co., Ltd. 73
Constables (Matlock Quarries) Ltd. ... 68
Contactor Switchgear, Ltd. 14
Conveyancer Fork Trucks Ltd. ... ... —
Coppee Co. (Gt. Britain) Ltd., The ... 40
Cox & Danks, Ltd. ... ... —
Crofts (Engineers) Ltd. 16
Crossley Bros., Ltd. 27

Davy & United Engineering Co., Ltd. 62
Delapena & Son Ltd. ... ... —
Dewhurst & Partner, Ltd. ... ... —
D.M.M. Machinery, Ltd. ... ... —
Dyson, J. & J. Ltd. 41

Efco-Edwards Vacuum Metallurgy Ltd. ... ... —
Electric Furnace Co., Ltd. ... ... —
Electric Resistance Furnace Co. Ltd. ... —
Electroflo Meters Co. Ltd. ... ... —
Elliott Bros. Ltd. ... ... —
English Electric Co., Ltd., The ... ... —
English Steel Corporation, Ltd. 2

Foster, Henry, & Co., Ltd. ... ... —
Foster Instrument Co. Ltd.... 54
Fraser & Chalmers ... ... —
Fuller's Earth Union Ltd., The ... ... —

General Electric Co., Ltd., The ... ... 26 & 55
General Refractories, Ltd. ... ... —
Gibbons Brothers, Ltd. ... ... —
Glenboig Union Fire Clay Co., Ltd. ... —
Green, E., & Son, Ltd. 72
Guest Keen Iron & Steel Co., Ltd. ... 32
Guest, Keen & Nettlefolds, Ltd. ... ... —
G.W.B. Furnaces, Ltd. 67

Hadfields, Ltd. ... ... —
Hallamshire Steel & File Co., Ltd., The ... —
Harland Engineering Co., Ltd. 49
Head, Wrightson & Co., Ltd. ... ... —
Head, Wrightson Machine Co., Ltd. ... 51
High Speed Steel Alloys Ltd. 66
Holmes, W. C. & Co., Ltd.... ... ... —
Holset Engineering Co., Ltd. 47
Honeywell-Brown, Ltd. 53
Howden, James & Co., Ltd. 23
Hudswell Clarke & Co., Ltd. 13
Hunslet Engine Co. Ltd. ... ... —
Huntington, Heberlein & Co. Ltd. ... —

Igranic Electric Co., Ltd. ... ... —
Ilford, Limited 44
Imperial Chemical Industries Ltd. ... ... —
International Construction Co., Ltd., The 29

Johnson's Rolls Ltd. ... ... —
Johnson Matthey & Co., Ltd. 33

Kestner Evaporator & Engineering Co., Ltd. ... 10
Kingscliffe Insulating Products, Ltd. ... —

Lamberton & Co., Ltd. Outside Back Cover
Lanarkshire Steel Co., Ltd., The 72
Lancashire Steel Corporation, Ltd., The ... 38
Lloyd, F. H., & Co., Ltd. 63
Lloyds (Burton) Ltd. ... ... —
Lodge-Cottrell, Ltd. ... ... —
Loewy Engineering Co. Ltd., The 8
London & Scandinavian Metallurgical Co. Ltd. 46
Lysaght's John. Scunthorpe Works Ltd. ... 75

Marshall, Thomas, & Co. (Loxley), Ltd. ... 11
Metropolitan Vickers Electrical Co., Ltd. —
Mobil Oil Co. Ltd. 35
Mitchell Engineering Ltd. ... ... —
Mond Nickel Co., Ltd. 57
Moorwood, John M., Ltd. ... ... —
Morgan Crucible Co., Ltd., The ... ... —
Morgan Refractories Ltd. 39
Morgardshammars Mekaniska Verkstads AB —
Moxey, Ltd. ... ... —

National Industrial Fuel Efficiency Service ... 68
Newalls Insulation Co.. Ltd. ... ... —
New Conveyor Co. Ltd. ... ... —
Newell, Ernest, & Co., Ltd. 60
Newton, Chambers & Co., Ltd. ... ... —

Oughtibridge Silica Firebrick Co. 17

Park Gate Iron & Steel Co., Ltd. ... ... —
Parsons, C. A. & Co., Ltd. ... ... —
Pearson, E. J. & J., Ltd. 52
Phillips Electrical Ltd. ... ... —
Pickford Holland & Co., Ltd. 77
Priest Furnaces, Ltd. 22

Rawlplug Co. Ltd., The ... ... —
Roberts, J. W. Ltd.... ... ... —
Robertson, W. H. A., & Co., Ltd.... ... 18

Salem Engineering Co., Ltd. ... ... —
Schloemann ... ... —
Shaw, John, & Sons (Salford) Ltd. 52
Shell-Mex & B.P., Ltd. 19
Sheppard & Sons, Ltd. 66
Simon-Carves, Ltd. 54
Smith, John (Keighley), Ltd. ... ... —
Smith Thomas & Sons (Rodley) Ltd. ... —
Smith & McLean, Ltd. 30
Stanton Ironworks Co., Ltd., The... ... —
Steel Co. of Wales, The ... ... 24 & 25
Stein & Atkinson, Ltd. ... ... —
Stein, John G., & Co., Ltd.... ... ... —
Stephens Silica Brick Co., Ltd. 36
Stewarts & Lloyds, Ltd. ... ... —
Stillite Products, Ltd. ... ... —
Sulzer Bros. (London) Ltd. 6
Summerson, Thos., & Sons, Ltd. ... ... —

Tattersall & Son Ltd. 76
Thermal Syndicate Ltd. 56
Thomas (Richard) & Baldwins, Ltd. ... —
Thornton, B., Ltd. 7

Union Carbide Ltd. 31
United Coke & Chemicals Co. Ltd.... ... —
United Steel Companies Ltd. 76

Vaughan, Edgar, & Co., Ltd. ... ... —
Visco Engineering Co., Ltd. 64

Wallsend Slipway and Engineering Co. Ltd. ... —
Ward, Thos., W., Ltd. 4
Warner & Co., Ltd.... ... ... —
Wellman Smith Owen Engineering Corporation Ltd., The 1
West Hunwick Silica & Firebrick Co., Ltd. —
Whitehead Iron & Steel Co., Ltd.... ... 28
Wild Barfield Electric Furnaces Ltd. ... —
Willan, G. L. Ltd. 34
Wincott, G. P., Ltd.... 15
Woodall-Duckham Construction Co., Ltd. 5

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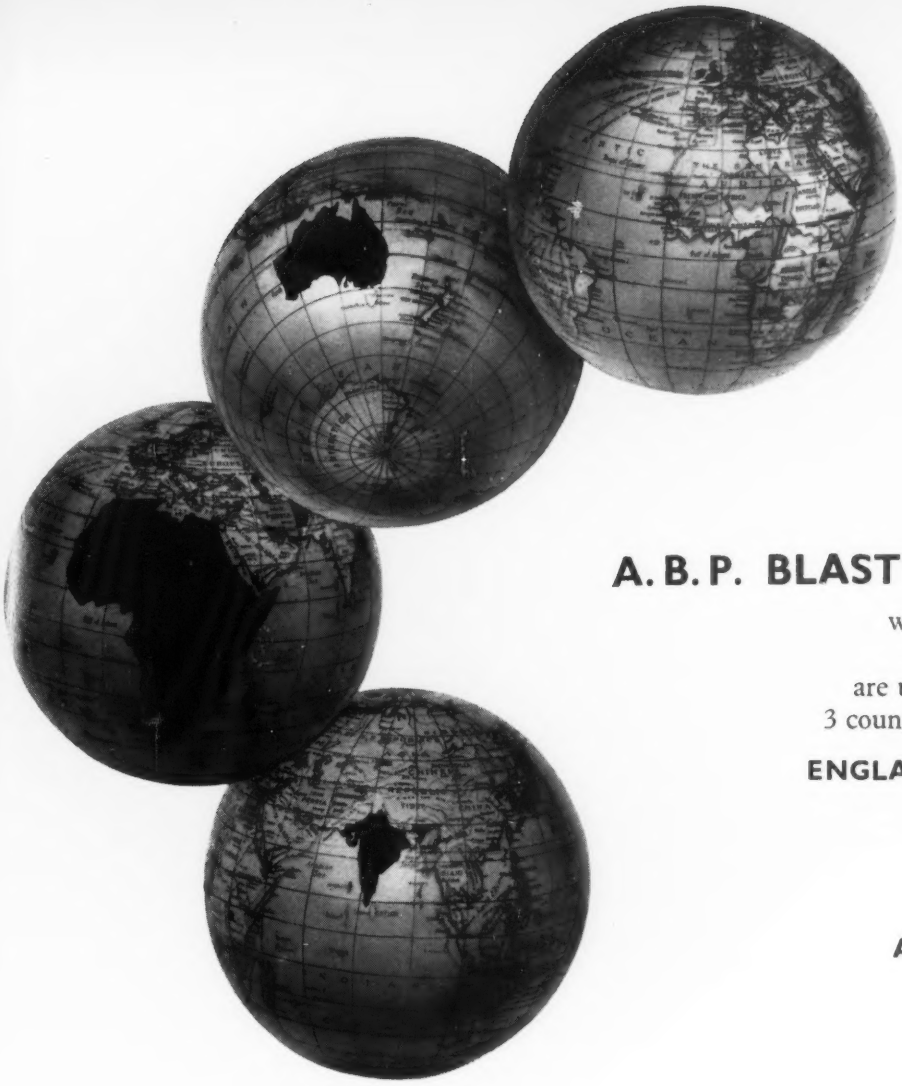

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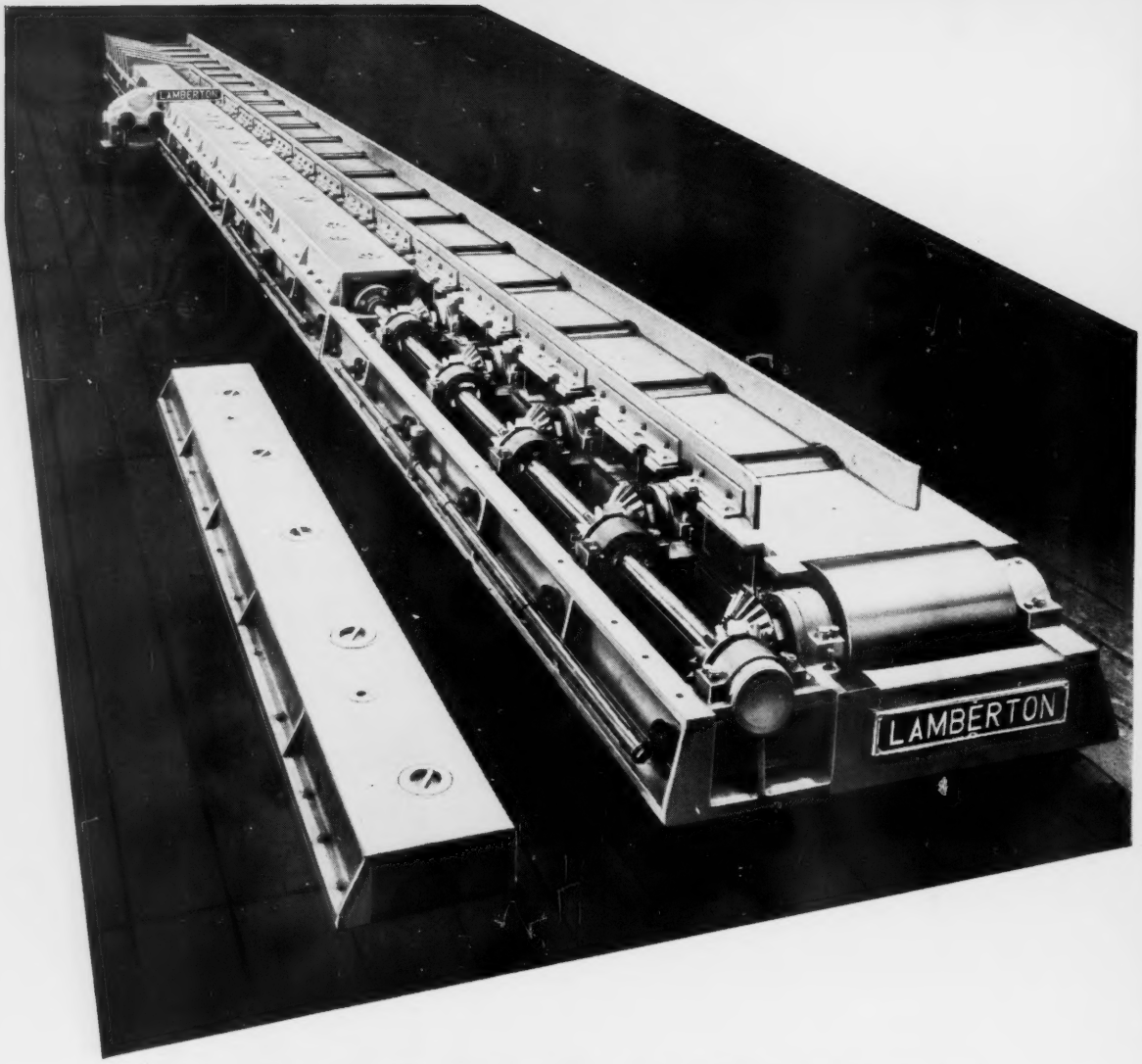

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